No. II 1935

THE JOURNAL

OF THE

IRON AND STEEL INSTITUTE

VOL. CXXXII.

EDITED BY

K. HEADLAM-MORLEY

SECRETARY

Reference
R669.106 I71 v.132
58765 6

PUBLISHED AT THE OFFICES OF THE INSTITUTE
28, VICTORIA STREET, LONDON, S.W.1.
1935

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PREFACE.

THE present volume contains eleven papers presented at the Autumn Meeting held in Manchester on September 16 to 18, 1935, together with the discussions and correspondence on them. The functions held and the works visited during the meeting are also reported. Bibliographical notes on the careers of deceased members are to be found at the end of Section I.

Two Special Reports were also presented at the Meeting in Manchester, namely, Special Report No. 9, "Sixth Report on the Heterogeneity of Steel Ingots," being a Report by a Joint Committee of the Iron and Steel Institute and the British Iron and Steel Federation to the Iron and Steel Industrial Research Council, and Special Report No. 10, "Waste-Heat Boilers in Open-Hearth Practice," the Second Report of the Open-Hearth Committee, being a Committee of the Iron and Steel Industrial Research Council. In view of the importance and interest of these two Reports, arrangements were made for them to be discussed at further meetings arranged jointly with a number of local institutions; accordingly, the entire discussion, both at the Manchester Meeting and at the further meetings, together with the correspondence and the Committee's Replies, have been held over and will be published later.

Section II. contains the usual notes on the progress of the home and foreign iron and steel industries as reported in the proceedings of scientific and technical societies and in the technical press, together with notices of new books presented to the Institute, and a bibliography of the principal works dealing with the metallurgy of iron and steel and allied subjects which have appeared during the half-year.

In front of the title-page is inserted a list of the British Standardised Steel Samples issued jointly by the Iron and Steel Institute and the National Physical Laboratory, showing where and on what terms the samples are available. Proposal forms for Membership and Associateship of the Institute will also be found in the same place.

28 VICTORIA STREET, LONDON, S.W.1.
December, 1935.

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Abbreviations Used in the Text.

A.	Ångström unit(s).
A.-C.	air-cooled.
a.c.	alternating current(s).
A.-H.	air-hardened.
amp.	ampère(s).
amp.-hr.	ampère-hour(s).
atm.	atmosphere(s) (pressure).
Bé.	Baumé (scale).
B.H.P.	brake horse-power.
B.O.T.	Board of Trade.
B.Th.U.	British thermal unit(s).
B.T.U.	Board of Trade unit(s).
B.W.G.	Birmingham wire-gauge.
C.	centigrade (scale).
cal.	calory ; calories.
c.c.	cubic centimetre(s).
c.d.	current density.
cg.	centigramme(s).
c.g.s.	centimetre - gramme - second unit(s).
cm.	centimetre(s).
coeff.	coefficient.
const.	constant(s).
c.p.	candle-power.
cu.	cubic.
cwt.	hundredweight(s).
d.c.	direct current(s).
dg.	decigramme(s).
diam.	diameter(s).
dm.	decimetre(s).
e.m.f.	electromotive force.
F.	Fahrenheit (scale).
ft.	foot ; feet.
ft.-lb.	foot-pound(s).
gal.	gallon(s).
grm.	gramme(s).
H.-F.	high-frequency.
H.P.	horse-power.
H.P.-hr.	horse-power-hour(s).
hr.	hour(s).
in.	inch ; inches.
in.-lb.	inch-pound(s).
K.	absolute temperature (scale).
kg.	kilogramme(s).
kg.-m.	kilogramme-metre(s).
km.	kilometre(s).
kva.	kilovolt-ampère(s).
kw.	kilowatt(s).
kw.-hr.	kilowatt-hour(s).
lb.	pound(s).
L.-F.	low-frequency.
m.	metre(s).
m.-amp.	milliampère(s).
m.-volt	millivolt(s).
max.	maximum.
mg.	milligramme(s).
min.	minimum ; minute(s).
ml.	millilitre(s).
mm.	millimetre(s).
m.m.f.	magnetomotive force.
N.	normal (solution).
N.T.P.	normal temperature and pressure.
O.-H.	open-hearth ; oil-hardened.
O.-Q.	oil-quenched.
oz.	ounce(s).
p.d.	potential difference.
pH	hydrogen-ion concentration.
R.	Réaumur (scale).
r.p.m.	revolutions per minute.
sec.	second(s).
sp. gr.	specific gravity.
sq.	square.
T.	tempered.
temp.	temperature.
v.	volt(s).
va.	volt-ampère(s).
w.-hr.	watt-hour(s).
W.-Q.	water-quenched.
yd.	yard(s).
°	degree(s).

SECTION I.

MINUTES OF PROCEEDINGS

AND PAPERS OF

THE IRON AND STEEL INSTITUTE.

AUTUMN MEETING

1935

Editorial assistance has been given by A. E. CHATTIN, B.Sc. (Hons. Met.), A.I.C., Assistant Secretary, in the preparation of this Section.

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MINUTES OF PROCEEDINGS

AND

PAPERS AND DISCUSSIONS

AT THE

AUTUMN MEETING IN MANCHESTER, 1935.

THE AUTUMN MEETING OF THE IRON AND STEEL INSTITUTE took place in Manchester during the week beginning September 16, 1935. Two sessions for the discussion of papers were held, the first on Tuesday, September 17, in the Great Hall of the Manchester College of Technology at 9.30 A.M., and the second in the Students' Common Room of the College on Wednesday, September 18, at 9 A.M. The Rt. Hon. The Earl of Crawford and Balcarres, P.C., K.T., President of the Manchester Reception Committee, took the Chair at the opening of the meeting, and was followed during the discussions by Sir Harold Carpenter, D.Sc., F.R.S., President of the Institute.

WELCOME TO THE INSTITUTE.

The Institute was welcomed by Lord Crawford and Balcarres and by the Rt. Hon. the Lord Mayor of Manchester (Alderman Samuel Woollam, J.P.).

The Rt. Hon. THE EARL OF CRAWFORD AND BALCARRES (President of the Manchester Reception Committee) said that this was the first occasion during the present century that the people of Lancashire had had the opportunity and the honour of welcoming the Iron and Steel Institute to their county. They did so with the utmost satisfaction. Not only were they closely concerned and intimately allied with all the great industries which the Institute represented, but the inhabitants of Lancashire, and particularly of the City of Manchester, were in the very centre of a great aggregation of population, of a wide and varied distribution of industry, of a very large and a very active market, and, he might add—and he hoped that, after their experiences the following day, the members would agree with him—of a great spirit of enterprise.

He was glad to welcome the members of the Institute in order to dispel one of two great fallacies which seemed to envelop British opinion. On one he would not dwell; indeed, he would not name it; but the second fallacy was the fallacy that Lancashire was down and out. It would, he hoped, be possible to convince the members that in many ways the people of Lancashire were playing their full part in the scrutiny and solution of all those difficult problems which faced the enterprise and the prosperity of the country. They, on their part, were grateful for and would be heartened by the presence in their midst of so many who represented the great industries of steel and of iron.

He had the honour to invite their most distinguished guest and host, the Lord Mayor of Manchester, to welcome the Institute to the City.

The Rt. Hon. THE LORD MAYOR OF MANCHESTER (Alderman Samuel Woollam, J.P.) said that it was a great privilege for him, as representing the citizens of Manchester, to offer to the Institute a very hearty welcome in the name of the citizens. Lord Crawford had very rightly drawn attention to the fact that a great many people who were not acquainted with Lancashire and Manchester had the idea that Lancashire was down and out. He could give the assurance that Lancashire, and Manchester in particular, was anything but down and out.

When the Chairman of the Reception Committee, Mr. J. E. James, approached him to see whether the Corporation would afford facilities for welcoming the Institute to Manchester, he readily assented and said that they would do everything possible to make the visit of the Institute a success. One of the most encouraging things during his year of office had been the great improvement in Manchester and in Lancashire generally of the engineering trades, which afforded some compensation for the loss of trade in the cotton industry. He believed that, as compared with a year ago, at least a thousand more people were employed by the engineering trades in Manchester. That was very encouraging, because as Chief Magistrate of the city he wished to see employment found for the people of Manchester, for employment created happy homes.

Judging by the programme which the Institute had arranged, every facility had been afforded for the members to see what Manchester could do. The ladies were to visit that morning the premises of firms who would show them some of the fabrics which Lancashire could produce—and Lancashire could produce some of the finest cotton and silk goods in the country. The shopkeepers of the city had laid themselves out to show their best goods, and he hoped the ladies of the party would do them

the honour of visiting their shops and seeing what they had to offer.

The College of Technology, in which the Institute was meeting, had done and was doing to day a great deal to assist the engineering works of the city and the adjoining area, and he would be failing in his duty if he did not say how glad the civic authorities were to see present that day the Principal of the College (Mr. B. Mouat Jones), who had been ill for some time and whose illness had given all those associated with the College a very anxious time. Mr. Mouat Jones had done admirable work in connection with the College, and it was a pleasure to see him about once more ; it was to be hoped that his health had been completely restored, so that he might continue his good work.

He would like to say, not only as Lord Mayor of Manchester, but as Chairman of the Education Committee, how glad he was that the Board of Education were now awake to the desirability of enquiring into the work of technical education in this country, because he felt that, although much had been done, much remained to do. He would like to appeal to employers of labour in the engineering industry to do whatever they could to help, so that when students were trained to enter the engineering industry employment could be found for them in and near the city.

The members of the Institute could rest assured that Manchester would not only do its duty in seeing that they had a good time during their visit, but was at all times ready to do its best to encourage the engineering trades, because by so doing employment was found for its citizens. If during the next few days the members became tired of looking at machinery, for relaxation they would be able to visit one of the most important agricultural shows in the country, now being held in the neighbourhood of Manchester. In the name of the City he offered them a very hearty welcome, and hoped they would carry away with them very pleasant recollections of their visit.

Sir HAROLD CARPENTER, D.Sc., F.R.S. (President of the Institute), who responded, thanked Lord Crawford and the Lord Mayor of Manchester, on behalf of the members of the Institute and their ladies, for the delightful welcome they had given them. Lord Crawford had referred to the fact that this was the first visit which the Institute had paid to Manchester during the present century. That was true, but he would like to point out that in the early days of the life of the Institute, at a time when it needed friends, there was no city which showed itself more friendly and generous than Manchester. The Institute was founded in 1869, and Manchester, although, as had been said, not pre-eminently a centre of metallurgy, showed its

enterprise in recognising the work that such an Institute could do for its interests, with the result that the Institute met three times in Manchester during the last century. There was no other city in the country which had that record.

The first meeting was in 1875, when the Institute was six years old, and its membership numbered only a few hundreds. The President just previous to that meeting was the famous Sir Lowthian Bell, but the President at that first meeting in Manchester was Mr. William Menelaus. On looking through the *Journal* which gave an account of that meeting, he had been interested to notice the famous names of Bessemer, Crawshay, and that famous citizen of Manchester, Sir Joseph Whitworth. The second visit was in 1887, the year of the Jubilee Exhibition, and on that occasion the President was a Manchester citizen, Daniel Adamson—a name well known and honoured in Manchester to-day. There was a very remarkable exhibit at that meeting by the firm of Sir Joseph Whitworth & Co. of hollow and solid ingots of cast steel made under their system of fluid compression, and hollow shafts, cubes and rings made under their system of hydraulic forging machines. One of the contributors to the discussion on that occasion was a Mr. F. W. Harbord, who was now one of the Past-Presidents of the Institute. He had hoped that Mr. Harbord would be present that morning; he was to be congratulated in that, although it was now forty-eight years since he first contributed to a discussion before the Institute, he was still in the full tide of his abilities.

The third visit was in 1899, and the President on that occasion was the late Sir William Roberts-Austen, a name of special interest to him personally, because Sir William was his predecessor at Manchester University and at the Royal School of Mines. Papers were presented at that meeting which had had a most profound effect on the engineering industry. It was the beginning of the period of metallography. One of the most important papers at that time was by Dr. Stansfield, on "The present position of the solution theory of carburised iron."

The present was therefore the fourth visit of the Institute to Manchester. It was of interest to mention that more steel had been produced since the last visit of the Institute to Manchester than in the whole of the previous history of the world. That might sound a very remarkable statement, but it was quite true. The development of steel metallurgy had been very remarkable, but he wanted to stress not the quantitative side of production, but the quality of the products, and he would like to give two or three examples showing how the advance in the industry which the Institute strove to serve had helped the industries in the Manchester district.

He would begin with cutting tools. When the Institute

last met in Manchester, high-speed steels did not exist. The whole of the revolution created by the discovery of the rapid cutting steels had taken place since that period. He would like to mention one of the early papers on that subject, a paper contributed by a citizen of Manchester, Mr. Leadhill. There were some present in the room that day who would remember that paper, which was a very important one. Then there were the high tensile steels which were used to-day in structural work. The most remarkable example of their use was the Sydney Harbour bridge, where 38,000 tons of steel were formed into a single arch. That would have been quite impossible thirty-five years ago. The heat-resisting and scale-resisting steels, which were so much used in plant to-day and which enabled high temperatures and pressures to be reached, had also been developed during this period; but probably the greatest discovery had been that of what was called stainless steel. The discovery of stainless steel was one of the major inventions of the century, and it could be said that the problem of the corrosion of steel was now solved. He meant, of course, the technical problem. Steel which would not corrode could be produced for a very large number of purposes, but whether people would pay to use it under such conditions was a commercial matter. Scientifically, however, the problem of corrosion had been solved, and to-day we were in the full tide of development of alloy steel metallurgy. The previous services of the Institute to industry had been great, but he thought that to-day they were greater than they had ever been.

Presentation to the Former Secretary of the Institute, Mr. G. C. Lloyd.

Sir Harold Carpenter (the President) said the members would recall that two years ago Mr. Lloyd had retired from the Secretaryship of the Institute, after having been Assistant Secretary from 1900 to 1904 and Secretary from 1909 to 1933. At the time of his retirement the Council made a presentation to him, but since then a movement had been initiated among the members themselves, who were desirous of not letting Mr. Lloyd go without taking their share in a presentation. That movement had the fullest support of the Council. Professor J. H. Andrew had been very active in the matter, and in a moment he would ask him to make the presentation. He wished to mention that the French members had not been invited to subscribe on this occasion, because they had, with that generous charm which was characteristic of their nation, already made a presentation to Mr. Lloyd immediately on his retirement; he would like also to add that a letter had been received from

German members and friends in which they conveyed their best wishes to Mr. Lloyd and expressed the feelings of esteem and regard in which they held him.

Professor J. H. ANDREW (Sheffield), in making the presentation, said he was sure that every member of the Institute would like to associate himself with the kind remarks which Sir Harold Carpenter had made with reference to Mr. Lloyd. They all recognised that the splendid position which the Institute had attained among the metallurgical societies of the world was due in no small measure to the efficient guidance which it had had at the hands of Mr. Lloyd, and they had always regarded him not only as their secretary, but as their counsellor and friend. His charming personality, his tact, and his modesty, emanating from a spirit of kindliness, had endeared him to them all.

They were presenting Mr. Lloyd that day with a desk and a cheque. Unfortunately, a desk was rather a large article of furniture to hand over on such an occasion, and he would therefore hand to Mr. Lloyd an inscribed plate which would be firmly fixed to the desk, on which the inscription read :

To
GEORGE CHRISTOPHER LLOYD
SECRETARY
OF THE IRON AND STEEL INSTITUTE
1909 TO 1933.
AS A TOKEN OF THEIR AFFECTION AND ESTEEM
FROM HIS FRIENDS
THE MEMBERS.

He ventured to suggest that no words were more sincerely engraved at any time.

Addressing Mr. Lloyd, Professor Andrew continued : "We ask you on behalf of the members of the Iron and Steel Institute to accept this as a token of their affection, and we sincerely trust that you and Mrs. Lloyd will enjoy long life and happiness. In other words, may you enjoy all the good things that you have so honourably and truly earned."

The presentation was then made amid prolonged applause.

MR. G. C. LLOYD, in reply, said it was indeed difficult for him to find words in which to express his gratitude for the very kind manifestation of their goodwill which the members had shown him. He would like to thank them each personally, but they would find that a tiring business. He had actually been associated with the Institute for twenty-nine years ; he

was Assistant Secretary for four years, and then, after a short interval, returned to the Institute to take up the office of Secretary, which he held for twenty-five years. That might not seem a very long time, but he started rather late in life. It was his own suggestion that the time had come when it was desirable to retire, for he had become sadly conscious of what metallurgists called the ageing effect. He felt that the time had come—as it must come to everyone—to bid farewell to what had been the chief interest of his life and to make room for someone more capable of bearing the increasing responsibilities imposed by the work of the Institute.

It had always been his aim to make himself as accessible as possible to the members, and it had been a very great privilege to find that he had gained a great multitude of friends in the course of his years of service. The great majority of his friends, of course, were among the British members, but he had also to express his gratitude to his French friends, who, a little while ago, in that charming and unobtrusive manner which was theirs, at a little gathering in London, presided over by Professor A. M. Portevin, had presented to him a precious token of their friendship and esteem. That was done privately, but he wished now to take the opportunity of thanking them publicly for it. He would also like to call to mind the German members, with whom he had always been on the very best of terms, and from whom he had received many manifestations of kindness. There were likewise numerous friends in Belgium, Italy, Sweden, Spain and Czechoslovakia in addition to the countries to which he had referred, just to mention those countries in which meetings had taken place during his period of office, and not forgetting those members overseas in the Dominions, in the United States of America, and in Japan. He had the most pleasant recollections of the many miles he had travelled by land and water in company with the President, Council and members of the Institute, and together they had tramped many a weary mile through innumerable iron and steel works, and other establishments of a less strictly professional interest.

He would like to thank all those concerned. The Iron and Steel Institute was a sort of League of Nations, and he wished to thank every one of its members for their kindness to him.

He felt he must add a few words on the subject of the staff. No man was ever better served by a more loyal and devoted staff, both men and women, than he had been. He need mention no names; they were still with the Institute, and the members knew them all personally. Finally, there was one other person to whom he owed the greatest debt of gratitude, and that was his wife, who was beside him in the meeting now. She wished to join with him in thanking the members for the very great

kindness which they had shown him that day. Without her loving care and strict supervision he doubted whether it would have been possible for him to be there at all.

The Lord Mayor of Manchester then left the meeting, and Lord Crawford vacated the Chair, which was taken by Sir Harold Carpenter, D.Sc., F.R.S., President of the Institute.

FORMAL BUSINESS.

The minutes of the previous meeting were taken as read and confirmed.

Ballot for Election of New Members.

Mr. H. W. SOWARD (London) and Mr. H. C. WOOD (London) were appointed scrutineers of the ballot, and they subsequently reported that the following fifty-one candidates for membership and seven candidates for associateship had been duly elected :

MEMBERS.

ALBERT, WERNER, D.Ing. (Aachen), Dipl.-Ing. (Munich) . . .	Landore, S.O., Glamorganshire.
BALLAY, MARCEL, Dr. Sc. . .	Paris, France.
BENBOW, W. E.	London.
BERESFORD-JONES, PAUL, B.A.(Cantab.)	Chester.
BROOME, LEONARD REEVES . .	Annan, Scotland.
BUZON, VICENTE	Santander, Spain.
CHAMEN, LAWRENCE HERBERT . .	London.
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DAVISON, VIVIAN WESTGARTH, B.Met. .	Barrow-in-Furness.
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GEARY, WILLIAM	Middlesbrough.
GERMEAU, EUGÈNE	Clabecq, Belgium.
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WRIGHT, JOHN TOM	West Hagley.
WRIGHT, R. A. M.	Bellshill, Lanarkshire.

ASSOCIATES.

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WARD, STACEY GEORGE, Ph.D., M.Sc., D.I.C., A.R.C.S.	Scunthorpe, Lincolnshire.
WELLBURN, EDGAR ROBERT, B.A., B.Sc.	Sheffield.
WILLIAMS, THEON EDWARD HOWARD	Scunthorpe, Lincolnshire.

Members of Council Retiring in 1936.

The SECRETARY, in accordance with Bye-Law 10, announced the following list of Members of Council due to retire in rotation

at the next Annual Meeting. *Vice-Presidents*: Mr. A. Hutchinson; Mr. John Craig, C.B.E.; Mr. H. Spence Thomas. *Members of Council*: The Rt. Hon. the Earl of Dudley, M.C.; Captain R. S. Hilton; Mr. I. F. L. Elliot; Mr. E. F. Law; Dr. T. Swinden.

Papers Presented.

The following papers were presented:

"Sixth Report on the Heterogeneity of Steel Ingots." Being a Report by a Joint Committee of the Iron and Steel Institute and the British Iron and Steel Federation to the Iron and Steel Industrial Research Council. (Special Report No. 9.)

"Waste-Heat Boilers in Open-Hearth Practice." Second Report of the Open-Hearth Committee, being a Committee of the Iron and Steel Industrial Research Council. (Special Report No. 10.)

E. W. Fell: "The Piobert Effect in Iron and Soft Steel."

E. T. Gill and R. Goodacre: "Some Aspects of the Fatigue Properties of Patented Steel Wires. II.—Note on the Effect of Low-Temperature Heat Treatment."

W. E. Goodrich: "The Penetration of Molten White Metals into Stressed Steels."

R. H. Greaves: "The Properties of Some Low-Nickel Steels containing Manganese."

R. Griffiths: "Subcutaneous Effects during the Scaling of Steel."

A. M. Herbert and F. C. Thompson: "The Use of the Hele-Shaw Apparatus in the Investigation of the Flow of Metals."

W. E. Hoare and B. Chalmers: "Examination of the Surface of Tinplate by an Optical Method."

C. H. M. Jenkins: "Behaviour of Mild Steel under Prolonged Stress at 300° C. Part II.—Experiments on Concentrated Stress in Notched and Drilled Specimens."

C. H. M. Jenkins and G. A. Mellor: "Investigation of the Behaviour of Metals under Deformation at High Temperature. Part I.—Structural Changes in Mild Steel and Commercial Iron during Creep."

E. Maurer and W. Bischof: "The Distribution of Phosphorus between Metal and Slag in the Basic Process of Steel Manufacture."

A. M. Portevin and R. Castro: "The Morphology of Inclusions in Siderurgical Products."

Votes of Thanks.

The President (Sir Harold Carpenter, D.Sc., F.R.S.) proposed a hearty vote of thanks to:

The Rt. Hon. the Lord Mayor of Manchester and the City Authorities, who could be assured, he said, that the members would take away with them very happy memories of Manchester.

The Principal and Council of the College of Technology, for permission to use the College for the meeting.

The Directors of the various works to which the members were making visits.

Lord Crawford, Mr. James and the Reception Committee, who had made such admirable arrangements for the entertainment of their visitors, and especially to Mr. H. D. Lloyd, the Hon. Secretary, on whom the chief burden had fallen.

No words of his were necessary to commend the resolution to the meeting; he felt sure it would be passed with acclamation.

Professor HENRY LOUIS, M.A., D.Sc. (Past-President), seconded the motion, which was carried with acclamation.

Mr. W. R. LYSAGHT, C.B.E. (Past-President), proposed a very cordial vote of thanks to the President, Sir Harold Carpenter, who had, he said, conducted the meetings in his usual sympathetic and charming way. He hoped Sir Harold would not feel too tired when he went home again; at least he could say that the Institute had had a very successful meeting.

The vote of thanks was carried with acclamation, and, the President having briefly acknowledged the compliment, the proceedings then terminated.

THE DISTRIBUTION OF PHOSPHORUS BETWEEN METAL AND SLAG IN THE BASIC PROCESS OF STEEL MANUFACTURE.*

By E. MAURER and W. BISCHOF.
(School of Mines, Freiberg in Saxony.)

Synopsis.

First, a series of melts in the laboratory is studied with a view to determining the distribution of phosphorus between metal and slag in the presence of pure iron-phosphate slags, the effects due to silica and to temperature conditions being also examined. In the light of these results the authors go on to consider the occurrence of dephosphorisation in the actual process of steel making, drawing attention to the importance of changes in the manganese content of the steel and examining a great number of operating analyses. A composite diagram is built up whereby the distribution of phosphorus may be inferred from the lime and silica contents of the slag. Further sections of the paper deal with the dependence of dephosphorisation in the open-hearth process on temperature, and with the influence of manganous oxide, magnesia and alumina. The paper closes with a summary.

I.—Introductory Remarks Concerning Phosphorus Reactions in Steel Making.

The basic process of steel making, with its lime slag formation, offers a means of removing phosphorus from the charge and of obtaining a steel of low phosphorus content. It is not found, however, that complete freedom from phosphorus can be secured, as a certain minimum amount of this element always remains in the steel, according to the conditions of slag formation, temperature, and practical working which obtain, besides which it is not unusual to observe what is known as "rephosphorisation" or transference of some of the phosphorus from the slag back into the steel. It is clear, then, that the division of phosphorus between the slag and the steel in the processes of manufacture can be referred to certain laws of equilibrium; in the laboratory various means have been worked out for proportioning the phosphorus accordingly, but hitherto the possibility of making a direct application of these researches in practice has been lacking.

It will be the purpose of the present paper to attempt an interpretation of analyses derived from practical operation and

* Received June 3, 1935.

to link them up with the experimental work described by the authors in an earlier paper,[1] in an endeavour by this means to determine the respective influence of the steel and the slag components on the distribution of phosphorus in steel manufacture by the basic process, in a similar way to that followed in the research on the manganese reaction, or in that on the sulphur content in open-hearth furnace gases,* also by the present authors.[1,2a,3]

Probably the phosphorus occurs in the steel as a phosphide of iron and manganese, for in view of its low boiling point it is impossible to maintain that it can be present to any appreciable extent as an element ; were this the case complete dephosphorisation of the steel by volatilisation of the phosphorus ought to be possible at those temperatures, in the region of 1,600° C., which attend the formation of steel. The high values of the heat of formation of both of the iron phosphides known at present—43 k.-cal. for Fe_3P and 60 k.-cal. for Fe_2P[5]—offer further arguments against the existence of elementary " free " phosphorus in the molten steel.

The oxidation of phosphorus in steel and the transfer of the product of oxidation to the slag is apparently to be visualised as follows : The phosphide is oxidised by dissolved ferrous oxide to P_2O_5, and at the same time the phosphorus pentoxide which is being formed combines to give an iron phosphate. It would appear that phosphoric acid, with its high vapour pressure, is (like the phosphorus) not present in the free state to any appreciable extent. The oxidation of phosphorus in steel can then be represented by the following equation :

$$2Fe_3P + 8FeO \rightleftarrows (FeO)_3\,P_2O_5 + 11\,Fe \quad . \quad . \quad . \quad . \quad . \quad (1)$$

but this is not to be read as implying a definite phosphide or phosphate. The reaction has also been expressed as follows :

$$2Fe_3P + 5FeO \rightleftarrows P_2O_5 + 11\,Fe \quad . \quad . \quad . \quad . \quad . \quad (2)$$

it being assumed that further combination of the P_2O_5 to form iron phosphate occurs at the surface of junction of the steel with the slag. But for the reasons given above it would appear incorrect to write :

$$2P + 5FeO \rightleftarrows P_2O_5 + 5Fe \quad . \quad . \quad . \quad . \quad . \quad . \quad (3)$$

It was found in the experimental research by W. Bischof and E. Maurer,[1] mentioned above, to be impossible to formulate the law of the mass relationships in accordance with equation (1) ; moreover, a constant of equilibrium according to equation (3)

* The authors take this opportunity of correcting an error in their former paper, " The behaviour of sulphur in open-hearth furnace gases," *Journal of the Iron and Steel Institute*, 1934, No. I. wherein on p. 142 the ordinate in Fig. 8 marked " % " should read " Grm. per cu. m."

would increase with the concentration more rapidly than is implied by the simple ratio $(P_2O_5)/P$, although, from the equation above, the constant would be more likely to agree with the quotient $(P_2O_5)/P^2$. This observation has been confirmed from another source,[4,6] but in the absence of any exact knowledge of the combinations arising in the reaction the arguments are irrelevant. H. Schackmann and W. Krings[4] have also put forward equation (3) as expressing the mass relationships, with P substituted for P^2; in this way it is true that closer conformity to actual conditions is obtained, but the foundation for equation (3) is quite inconsistent with the view taken by Schackmann and Krings, whose argument makes the mass relationships entirely arbitrary.

The mass law adopted by H.O. von Samson-Himmellstjerna,[7] agreeing with equation (1), likewise offers no advantage over the simple expression $(P_2O_5)/P$, apart from its practical utility being severely impaired by the high powers involved.

It is evident, then, that in the presence of pure iron phosphate slag no attempts to control definite combinations of the reacting substances have resulted in any consistent results, even in the simplest reactions. With slags containing lime or silica the case is obviously still more difficult, and under practical conditions insoluble.

The phosphate of iron formed in accordance with equation (1) passes into the slag, where, in the presence of lime, it is converted to a more or less stable phosphate of calcium. According to the form of the latter these reactions may be written:

$$\text{either, } (FeO)_3P_2O_5 + 3CaO \rightleftarrows (CaO)_3P_2O_5 + 3FeO \quad . \quad . \quad (4a)$$

$$\text{or } (FeO)_3P_2O_5 + 4CaO \rightleftarrows (CaO)_4P_2O_5 + 3FeO \quad . \quad . \quad . \quad (4b)$$

On combining these equations with (1) we obtain:

$$2Fe_3P + 5FeO + 3CaO \rightleftarrows (CaO)_3P_2O_5 + 11Fe \quad . \quad . \quad (5a)$$

or, alternatively,

$$2Fe_3P + 5FeO + 4CaO \rightleftarrows (CaO)_4P_2O_5 + 11Fe \quad . \quad . \quad . \quad (5b)$$

The following mass relationship emerges:

$$\frac{P^2\,(FeO)^5\,(CaO)^3}{(P_2O_5)} = K \quad . \quad . \quad . \quad (6)$$

the concentration of the phosphorus pentoxide being substituted for that of the calcium phosphate, and the concentration of phosphorus as the element itself for that of the Fe_3P. This expression of the mass law—somewhat altered by C. H. Herty[8] and later by C. Schwarz, E. Schröder and G. Leiber on account of a different definition of (CaO)—is not consistent with the present authors' earlier experimental work[1] (*loc. cit.* p. 419). H. Schenck[6] has attempted to reconcile expression (6) with actual conditions by the substitution of P for P^2 and by the introduction

of a correcting term ; the result, however, is that the mass law loses all relationship with the reaction (5) postulated at the beginning.

Moreover, it is impossible to refer to formula (6) as a law of mass action in the ordinary sense, since the lime occurring therein is actually present in one phase only. All that can be said is that the constants of mass relationship appropriate to a system having phosphorus-bearing iron for its metallic phase and phosphorus-pentoxide-bearing iron oxide for its slag phase, formulated in accordance with reactions (1) to (3) above, are subject to change in the presence of lime or silica.

The upshot of all this is as follows : It is certainly possible to attempt to express what happens in the phosphorus reactions by representing the experimental observations by an arbitrary formula which can be made to agree more or less with a postulated law of mass relationship ; but in the absence of exact knowledge as to the forms taken by the reacting substances the attempt only leads to further complication.

In the present work, then, the law of mass relationship will be replaced by the simple ratio :

$$V_P = (P_2O_5)/P$$

as already used for the first time by the authors in their earlier work,[1] and an endeavour will be made to establish this quantity for the basic process of steel manufacture, account being taken of the composition of the slag and of the metal.

II.—Melting Experiments by the Authors.

A knowledge of the distribution of phosphorus between iron on the one hand, and phosphate slag (free from lime and silica) on the other, is fundamental to the study of phosphorus reactions in steel making. This being the case, a series of additional laboratory experiments were carried out to check the authors' earlier work[1] in which doubt had been cast on the straight-line relationship of P and P_2O_5 (for phosphorus contents up to about 4 per cent.) claimed by Schackmann and Krings.[4]

(*a*) *Experimental Procedure.*—The following were the materials used in the smelting experiments :

(1) Iron phosphate (Kahlbaum) with 39·61 per cent. of Fe_2O_3, 19·63 per cent. of FeO, 24·61 per cent. of P_2O_5 and 16·11 per cent. of moisture.

(2) Ferrous oxide (Merck) with 43·08 per cent. of FeO, 54·61 per cent. of Fe_2O_3, and 1·89 per cent. ignition loss *in vacuo*. For the purpose of the experiments themselves the proportion of iron oxide is irrelevant, as iron oxide is not stable when melted in the presence of iron.

(3) Iron (*ferrum reductum* of Merck) with 0·2 per cent. of silicon.

(4) Ferro-phosphorus (Riedel de Haen) with 25·83 per cent. of phosphorus, 69·25 per cent. of iron, 4·75 per cent. of silicon, and traces of carbon.

(5) Silica (Kahlbaum).

The melting was done in crucibles of hard-glazed beryllium oxide, known from previous experience to be very resistant to slag formation from ferrous oxide and phosphate of iron, especially in the presence of silica.

The procedure followed in the experiments was very simple. The melting crucible, being relatively small, was surrounded by a tubular crucible of magnesia, preventing the access of furnace gases to the contents, and was placed in the Tammann furnace. The temperature was measured by optical means using a hot-wire pyrometer continuously checked against a thermocouple and adapted to be fixed almost directly below the bottom of the melting crucible ; in this way temperatures could be read accurately to 10° C. Heating to the temperature of the experiment took one hour and the subsequent rate of cooling down to 1,100° C. was about 120° per min. ; previous experience had shown that in this way the conditions for proper melting would be fully controlled.

A few observations arise regarding the chemical examination of the slag. In most cases the latter was found caked on the walls of the crucible, from which it had to be loosened mechanically ; in so doing some beryllium oxide was apt to come away with it, but this, on dissolving, remained behind as residue and could be filtered off. For analysis the finely powdered slags were dissolved in concentrated hydrochloric acid and nitric acid, the determination of silica being then effected by repeated evaporation of the hydrochloric acid solution, filtering of the residue, and heating. Next, the solution was saturated with sulphuretted hydrogen gas, rendered ammoniacal and then filtered, in order to determine the iron. The filtrate was evaporated to determine the P_2O_5, the ammonium salts being volatilised off and the residue absorbed in nitric acid ; the phosphoric acid was removed from the solution by ammonium molybdate and estimated in the usual way as magnesium pyrophosphate. Similarly, the phosphorus content of the metal was determined as magnesium pyrophosphate after dissolving in nitric acid.

No special determination was made of the iron oxide, as the amount present in the cold slag submitted to analysis would not correspond with that in the molten condition ; instead, the total iron content was determined and was re-calculated as ferrous oxide, the latter being in all probability the only oxide of iron that would be stable in the presence of iron itself at the high temperatures obtaining in the experiment.

(*b*) *Experimental Results and Conclusions in Regard to Pure Iron Phosphate Slags.*—Table I. contains a series of melts to which the addition was made of certain mixtures of pure iron powder with iron phosphate, or of ferro-phosphorus with iron and ferrous oxide. The melting period between 1,540° and 1,550° lasted up to 30 min. As appears from Fig. 1, the mean distribution of phosphorus agrees with the condition $V_P = 3 \cdot 55$ for up to 6 per cent. of phosphorus in the iron, but above this 6 per cent. the proportion appears to fall off in conformity with the authors' previous determinations, as shown by the lightly drawn curve.

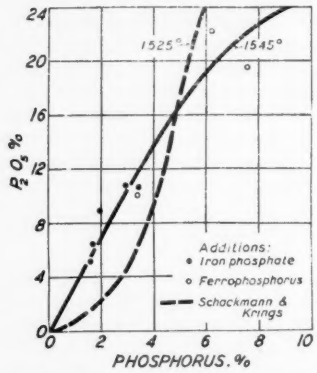


FIG. 1.—Distribution of phosphorus between iron and pure iron phosphate slags.

It is not immediately possible to compare these results with the curve by Schackmann and Krings shown by the broken line. The analyses of these two authors reveal a marked degree of impurity in the slags due to the clay of the melting crucible, and their argument that these impurities are without effect is not confirmed by a later check. It is already known from the researches of G. Mars[10] on practical Siemens-Martin melts and on electrical melts that dephosphorisation may be retarded by a high alumina content of the slag; the same would appear to be true of the melts carried out by Schackmann and Krings, for if the $(P_2O_5)/P$ ratio at 1,525° or 1,450°, as calculated from their analyses, be related to the alumina content the result is as shown in Figs. 2A and 2B, clearly revealing the influence of the alumina, and allowing the ratios 4·3 and 7·2 to be extrapolated for the case of slags free from alumina.

Schackmann and Krings, further, carried out a number of melts with no appreciable alumina content or other impurities, the results of which, between 1,350° and 1,370°, are plotted on

TABLE I.—*Melting Tests with Pure Iron Phosphate Slags.*

(1)	(2)	(3)	(4)	(5)	(6)	(7)	(8)	(9)	(10)	(11)	(12)
Melt No.	Additions in Grammes.				Duration of Melting. Min.	Temperature of Experiment. °C.	Phosphorus in the Metal. %	Slag Analyses.			P_2O_5/P
	Iron.	Ferro-phosphorus.	Iron Phosphate.	Ferrous Oxide.				P_2O_5. %	FeO. %	SiO_2. %	
22	12·90	..	2·32	..	20	1535	2·95	10·86	88·71	0·36	3·68
23	11·04	..	5·31	..	22	1540	3·40	10·72	87·31	2·29	3·15
24	10·75	..	1·29	10·75	30	1550	1·51	5·16	91·70	2·32	3·43
25	11·00	..	3·04	3·25	21	1540	1·98	8·98	89·20	0·68	4·75
26	5·24	..	2·60	6·96	15	1550	1·69	6·54	92·12	0·75	3·87
27	13·26	5·63	..	4·76	20	1530	6·14	22·24	75·28	1·18	3·63
28	7·60	7·80	..	6·99	18	1540	7·56	19·57	79·04	1·02	2·59
29	9·49	3·99	..	8·49	20	1550	3·34	10·10	87·24	0·95	3·00

TABLE II.—*Melting Tests with Iron Phosphate Slags Containing Silica.*

(1)	(2)	(3)	(4)	(5)	(6)	(7)	(8)	(9)	(10)	(11)	(12)	(13)
Melt No.	Additions in Grammes.					Duration of Melting. Min.	Temperature of Experiment. °C.	Phosphorus in the Metal. %	Slag Analyses.			*P_2O_5/P
	Iron.	Ferro-phosphorus.	Iron Phosphate.	Ferrous Oxide.	Silica.				P_2O_5. %	FeO. %	SiO_2. %	
30	4·49	2·88	..	3·07	2·52	18	1540	10·53	7·06	57·84	33·97	0·67
31	9·11	1·95	..	4·07	3·20	10	1550	4·68	3·46	66·00	30·22	0·73
32	7·50	5·05	..	6·57	1·26	10	1540	6·65	7·97	74·81	15·43	1·20
33	7·51	..	3·01	7·20	5·76	20	1540	3·61	4·49	71·18	22·31	1·24
34	5·75	..	4·75	3·25	2·58	10	1550	5·30	9·51	78·83	9·76	1·84
35	9·14	..	1·51	7·88	1·49	12	1540	7·86	9·97	71·35	17·45	1·12
36	8·30	1·61	..	6·93	1·20	15	1505	2·15	6·01	78·47	13·82	2·80
37	9·22	1·79	..	4·11	2·80	15	1510	3·44	3·10	64·50	30·91	0·90
38	6·90	..	3·56	7·10	5·52	18	1500	4·20	5·26	71·17	22·00	1·25

the ordinates (P_2O_5) and P in Fig. 3, together with those obtained in the present work between 1,540° and 1,550° and with those from the present authors' earlier work[1] at about 1,500°. All three of these curves at first ascend in straight lines and then bend over as a higher phosphorus content in the iron is reached, tending

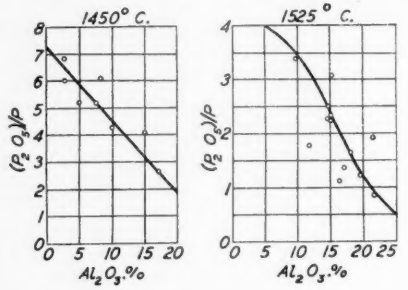


FIG. 2A. FIG. 2B.

FIGS. 2A and 2B.—Distribution of phosphorus between iron and iron phosphate slags in relation to alumina content using the experimental results of Schackmann and Krings.

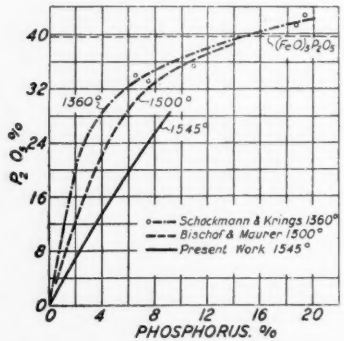


FIG. 3.—Distribution of phosphorus between iron and pure iron phosphate slag, and the effect of temperature thereon.

towards a limit which is probably represented by the P_2O_5 content of pure iron phosphate $(FeO)_3P_2O_5$. Thus, when the amount of phosphorus present is high the ratio $\frac{P_2O_5}{P}$ falls off sharply, and the lower the temperature of the reaction the more pronounced is this effect.

Working from analyses of steel and slag in the Lancaster process, it had been shown by the authors[1] that the rectilinear relationship between P_2O_5 and phosphorus holds good even when very small quantities are present and in practical melting operations.

With pure iron phosphate slags and low phosphorus contents, then, the ratio $(P_2O_5)/P$ is related to the temperature according to Fig. 4. As the temperature rises the ratio decreases; that is to say, dephosphorisation is retarded.

(c) *Experimental Results for Iron Phosphate Slags with Siliceous Additions.*—The results shown in Table II. were obtained from a series of melting tests in which siliceous additions were made to the slag. The temperature of the experiments was from 1,500°

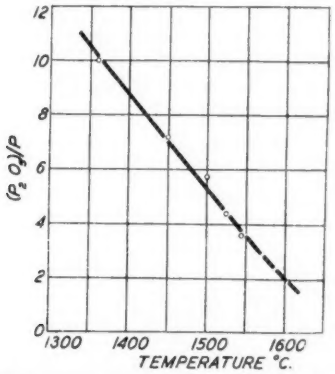


FIG. 4.—Effect of temperature on phosphorus distribution in the presence of pure iron phosphate slags.

to 1,510°, and from 1,530° to 1,550°, and their duration 10 to 20 min. The same proportionate distribution was obtained from both sides by using ferro-phosphorus and iron phosphate in the mix respectively. Fig. 5 again shows the variation of the ratio $(P_2O_5)/P$ with the silica content, dephosphorisation being greatly reduced by the latter; thus with 30 per cent. SiO_2 at 1,545° the dephosphorisation is only one-fifth of that at zero SiO_2 content. It is worth noticing also that the effect of a very high phosphorus content in the iron, as observed in the presence of pure iron phosphate slags, is not discernible as a further reduction of the ratio $(P_2O_5)/P$ in the case if the silica contents here used.

The temperature relationship revealed by both the curves appears to be somewhat less marked for higher silica contents.

III.—Distribution of Phosphorus Between Steel and Slag in Steel Making.

In order to arrive at the laws governing the distribution of phosphorus in practical operation an investigation was undertaken (as already mentioned) on the same lines as those used for determining the manganese reactions between the bath and the slag,[9] which, according to the most recent publications,[6] (*loc. cit.* p. 96, &c.) have been shown to lead to results of the highest accuracy and free from any objection.

What follows is based on these assumptions : The ratio $(P_2O_5)/P$ in steel making is assumed to tend to adjust itself always in the direction of equilibrium ; hence if samples are taken from the steel and the slag before and after a given interval of time the respective changes in the ratio $(P_2O_5)/P$ will show the direction

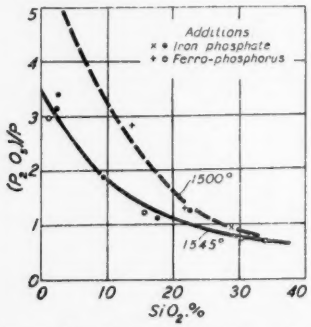


Fig. 5.—Relation between phosphorus distribution and silica content of iron phosphate slags.

in which equilibrium is to be sought. This, however, is on condition that no additions of ore, lime, &c., or metallic materials have been made in the interval ; also that no very pronounced change has occurred meanwhile in the other components of the slag.

Groups of analyses of this kind, susceptible of use in determining the distribution of phosphorus, are available in the abundant numerical data of Maurer and Bischof,[2] while O. Scheiblich's[11] analyses relating to the Thomas process have also been drawn upon. It is necessary, however, in making use of such data to exclude all samples in which it is obvious that the phosphorus content has been determined only by rough approximation, and to take account only of those in which it has been accurately fixed to within one thousandth of one per cent. The mean temperature is again, as in Maurer and Bischof's

work, taken as 1,600° C, corresponding to that which may be expected in the final period of the Siemens-Martin process; the assumption, as advocated in another quarter,[12] of a higher mean temperature than this is not adopted, as the new point of view on which that advocacy is based has not been free from criticism.*

(*a*) *Effect of Manganese on the Phosphorus Distribution*—Maurer and Bischof[2] (*loc. cit.* p. 148), having observed that the phosphorus content in steel had an appreciable influence on the manganese content, and this observation being later confirmed by H. Schenck[6] (*loc. cit.* p. 108) and O. Scheiblich,[11] the probability arises that the converse also is true and that manganese

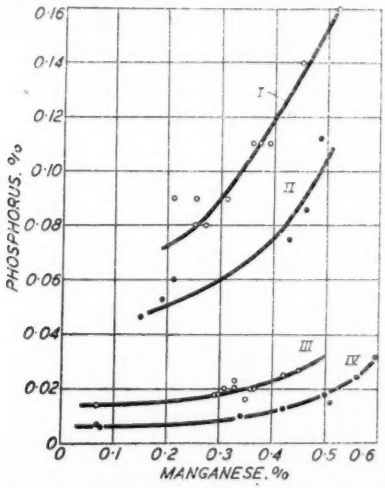


FIG. 6.—Relation between manganese and phosphorus contents with slags of otherwise unaltered composition.

influences the content of phosphorus. This would be explicable either on the assumption of a different mode of formation of the phosphides of iron and manganese, or on the supposition that they are both formed in the same way with different heats of formation. Schackmann and Krings mention briefly that manganese greatly strengthens the reducibility of P_2O_5 from the slag, but give no melting tests or numerical proofs in support of their contention, and H. Schenck[6] (*loc. cit.* p. 159), on the other hand, does not attribute to the manganese content of the steel any influence over the phosphorus distribution. If, however, it is actually the case that manganese influences the phosphorus

* See discussion on Schröder, [12] pp. 881–884, and also on Schenck, [6] pp. 14–15.

TABLE III.—*Phosphorus Content of Steel in Relation to Manganese Content.*

(1)	(2)	(3)	(4)	(5)	(6)	(7)	(8)	(9)	(10)
No.	Analyses of Steel.		Slag Analyses.						
	P. %	Mn. %	Fe. %	Mn. %	CaO. %	MgO. %	SiO_2. %	P_2O_5. %	Al_2O_3. %
I. 1	0·09	0·21	6·10	2·48	48·10	4·98	5·44	28·53	0·95
2	0·08	0·25	6·00	3·59	47·10	4·86	5·20	28·30	0·80
3	0·08	0·27	7·80	3·86	45·20	5·62	6·12	25·52	0·90
4	0·09	0·25	6·60	3·69	47·25	4·26	5·30	28·10	0·80
5	0·09	0·31	7·80	4·25	47·60	5·02	4·00	26·34	0·80
6	0·11	0·36	7·50	5·34	46·40	5·11	5·04	25·14	0·90
7	0·11	0·37	7·50	5·52	47·84	3·32	3·84	26·13	1·00
8	0·11	0·29	7·70	5·69	45·80	5·18	4·28	25·67	0·82
9	0·14	0·45	6·90	5·72	46·40	5·36	4·06	26·00	0·89
10	0·16	0·52	6·50	5·02	49·00	2·11	6·14	26·00	0·72
II. 1	0·046	0·15	9·53	3·44	48·43	..	7·02	19·97	..
2	0·053	0·19	7·71	3·98	49·37	..	7·72	20·88	..
3	0·060	0·21	7·86	3·78	48·42	..	6·27	21·85	..
4	0·075	0·43	6·62	5·50	54·39	..	6·04	23·10	0·17
5	0·086	0·46	6·05	5·25	55·84	1·52	6·76	22·43	0·29
6	0·112	0·48	6·00	5·12	54·88	1·49	6·42	23·45	0·28
7	0·172	0·57	7·49	4·73	55·07	..	6·46	23·68	0·29
III. 1	0·014	0·069	18·06	4·15	37·45	9·83	19·48	1·84	2·25
2	0·016	0·35	11·85	13·31	39·0	5·0	18·0	1·5	2·9
3	0·018	0·29	10·98	10·94	39·8	7·5	19·3	1·9	3·2
4	0·018	0·29	11·48	11·08	38·1	7·9	19·8	1·9	2·1
5	0·020	0·31	11·45	10·38	38·8	7·6	19·0	1·8	2·0
6	0·020	0·36	12·00	13·48	40·9	3·2	17·3	2·2	2·3
7	0·020	0·36	10·23	10·69	42·0	4·2	19·8	2·4	3·0
8	0·021	0·33	11·86	14·26	38·1	3·9	18·0	2·3	3·0
9	0·023	0·33	10·14	11·39	40·3	4·5	20·0	2·4	3·2
10	0·025	0·42	11·46	10·00	40·4	4·7	19·2	2·2	2·7
11	0·027	0·45	9·97	14·10	40·2	4·6	18·9	2·2	2·2
IV. 1	0·006	0·074	17·19	3·57	44·1	6·90	15·24	1·37	..
2	0·007	0·069	16·35	3·22	44·4	7·16	15·30	1·5	..
3	0·010	0·34	13·23	11·09	43·0	3·5	15·6	2·0	2·6
4	0·013	0·42	12·40	10·30	45·7	3·9	15·4	1·8	2·5
5	0·015	0·51	10·11	12·94	43·1	5·4	15·7	2·0	2·7
6	0·018	0·50	9·91	12·70	43·4	6·3	15·7	1·9	2·7
7	0·025	0·56	7·60	12·63	45·2	8·0	14·9	2·3	1·2
8	0·032	0·59	8·81	12·78	43·1	7·8	15·8	2·6	1·5

reactions, the fact cannot fail to emerge from a study of the numerical data mentioned above.

Table III. contains several groups of analyses of the steel and slag in the Thomas and Siemens-Martin processes. Since the slag analyses show good agreement in relation to CaO, SiO_2 and P_2O_5 between all the groups, the steel analyses ought (at any rate on the assumption of equal temperatures at the time of sampling) to show equal phosphorus contents; but instead of this being the case the relation of the phosphorus content to the manganese content is found to be that shown in the curves in Fig. 6. As manganese is added, the phosphorus in the steel

increases—at first slowly and then, corresponding to the higher manganese contents, rapidly. In order to examine the effect of the slag components on the phosphorus distribution it is necessary, therefore, first to establish how much phosphorus would be present in the samples of steel when the manganese content was nil regardless of the composition of the slag, and Fig. 7 has been

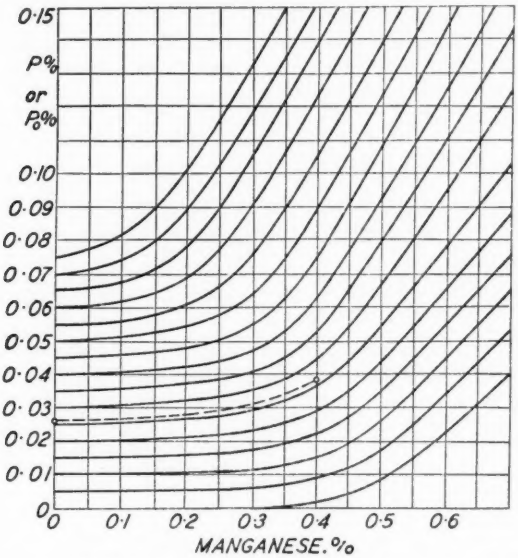


FIG. 7.—Diagram for determining the theoretical phosphorus content P_0 corresponding to Mn = 0 per cent. from the phosphorus and manganese contents found by analysis of the steel.

plotted for this purpose by interpolation from Fig. 6. The phosphorus corresponding to zero manganese in the steel when there is no change in the composition of the slag can be found by entering whichever curve corresponds to a known pair of values of the phosphorus and manganese contents (*e.g.*, 0·038 per cent. phosphorus and 0·4 per cent. manganese) and running down it to 0 per cent. Mn (*viz.*, 0·026 per cent. phosphorus). The phosphorus content found in this way will be denoted below by P_0 and the corresponding ratio $(P_2O_5)/P_0$ will be called V_{P_0}.

(*b*) *Effect of Lime and Silica on Phosphorus Distribution in Practical Steel Making.*—Figs. 8 to 14 show the results of an examination of analyses from the Siemens-Martin process based on the numerical data of Maurer and Bischof.[2] First P_0 was

determined, then V_{P_0} in relation to the silica content of the slag within a narrow range of lime contents; the direction in which V_{P_0} changes towards equilibrium as indicated by the analysis of a second sample is denoted by arrows.

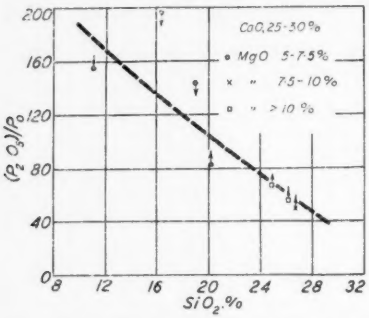


FIG. 8.—Effect of silica on the phosphorus distribution with 25 to 30 per cent. CaO.

In Fig. 8, by reason of the limited number and wide spacing of the measurements, the equilibrium corresponding to lime contents of 25 to 30 per cent. is not very certainly defined, and the same applies to lime contents of 30 to 35 per cent. in Fig. 9;

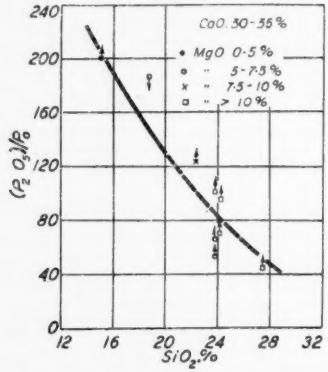


FIG. 9.—Effect of silica on the phosphorus distribution with 30 to 35 per cent. CaO.

both of these cases have reference to samples containing under 15 per cent. of magnesia. The pronounced increase in V_{P_0} due to lime is counteracted by silica. The points have been differently

marked according to the amount of magnesia in the samples; it is not possible, however, to establish that magnesia influences the results except to the extent that in Fig. 9 three groups of samples containing over 7·5 per cent. of magnesia show a slight tendency towards higher values of V_{P_0}.

A larger number of samples were available with lime contents in excess of 35 per cent., enabling them to be classified under lime contents of 35 to 37·5 per cent., 37·5 to 40 per cent., &c., and here again the points have been differently marked according as the MgO content was under 5 per cent., from 5 to 7·5 per cent., or from 7·5 to 10 per cent. All the samples containing from 35 to 37·5 per cent. of lime are entered in Fig. 10, in which the curve

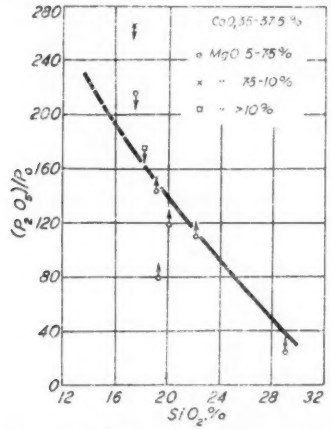


FIG. 10.—Effect of silica on the phosphorus distribution with 35 to 37·5 per cent. CaO.

of $(P_2O_5)/P_0$ plotted against the silica content is more definitely established; it cannot be said that the scattering of the points is such that the arrows do not indicate the direction of the appropriate curve. No consistent effect of magnesia up to 10 per cent. can be discerned.

Fig. 11 shows the condition with 37·5 to 40 per cent. of lime; despite the large number of samples (52 analyses in all), the values are not much scattered in relation to the curve as drawn. The points without arrows refer to samples which showed no further change in the ratio $(P_2O_5)/P_0$ during the progress of the melting. No influence of magnesia up to 10 per cent. can be established. The reversal of the curve is a noticeable feature, perhaps attributable to a change from the orthosilicate of lime

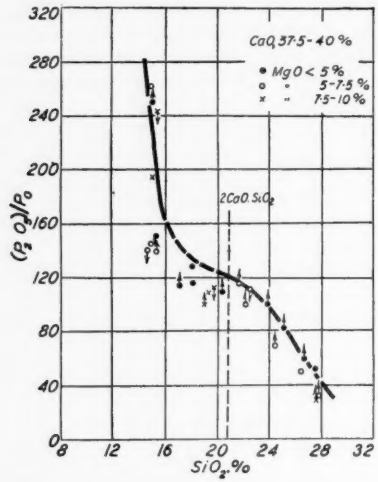


FIG. 11.—Effect of silica on the phosphorus distribution with 37·5 to 40 per cent CaO.

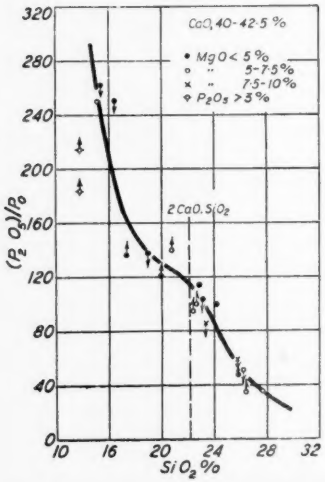


FIG. 12.—Effect of silica on the phosphorus distribution with 40 to 42·5 per cent CaO.

($2CaO.SiO_2$) to the metasilicate ($CaO.SiO_2$), the trace of the former being indicated by a broken line. Up to about 18 or 20 per cent. of silica the ratio decreases very rapidly ; it then becomes flatter until the lime content is completely saturated by the silica as $2CaO.SiO_2$, after which it again steepens.

Results for 40 to 42·5 per cent. of lime can be found from Fig. 12, in which the concentration of points enables the trend of the curve to be very accurately defined. Two groups of samples having over 3 per cent. of P_2O_5 show good agreement with those having a normal content of less than 2·5 per cent. of phosphorus pentoxide. Here again no influence is detectable of magnesia up to 10 per cent.

Fig. 13 shows the curve with lime contents of 42·5 to 45 per cent., to which the same remarks apply as to Fig. 12. Finally,

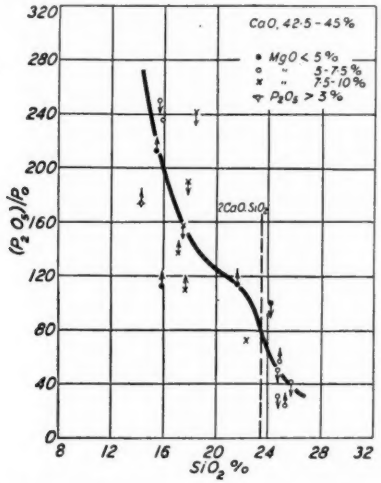


FIG. 13.—Effect of silica on the phosphorus distribution with 42·5 to 45 per cent. CaO.

Fig. 14 gives the results for 45 to 50 per cent. in a curve which lies a little lower than that for lime under 45 per cent. ; dephosphorisation, therefore, is again opposed by a high lime content.

The subsequent curves, Figs. 15 to 18, are studies of samples containing very little silica, obtained chiefly from the Thomas process, together with a few from the pig-ore process carried out in the Siemens-Martin open-hearth furnace. As most of the samples were taken individually or are final ones from the Thomas

process, it is hardly possible in any one of them to establish the direction of the reaction ; hence the curves could not be fixed by convergence and it was necessary to plot for mean values

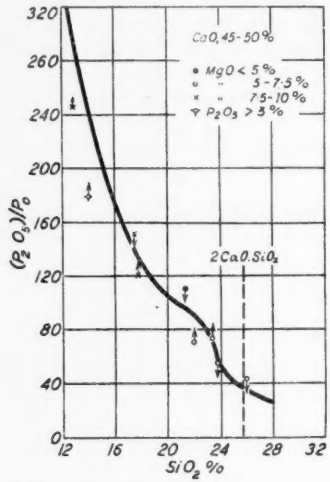


FIG. 14.—Distribution of phosphorus with slags containing 45 to 50 per cent. CaO.

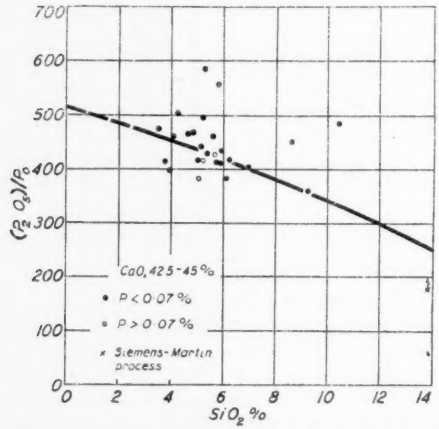


FIG. 15.—Phosphorus distribution with low-silica slags containing 42·5 to 45 per cent. CaO.

covering a small range of silica contents and checked by the indications for higher silica contents in Figs. 8 to 14 above. In this process of averaging, the most scattered of the values have been left out of account.

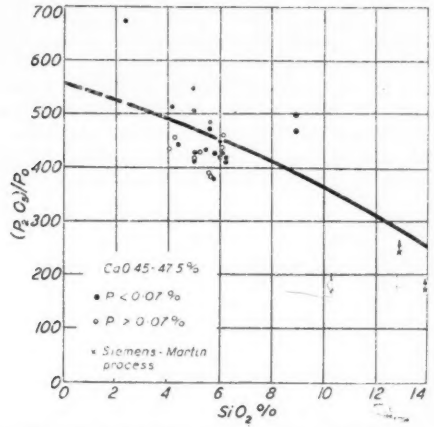


FIG. 16.—Distribution of phosphorus with slags of low silica content and containing 45 to 47·5 per cent. CaO.

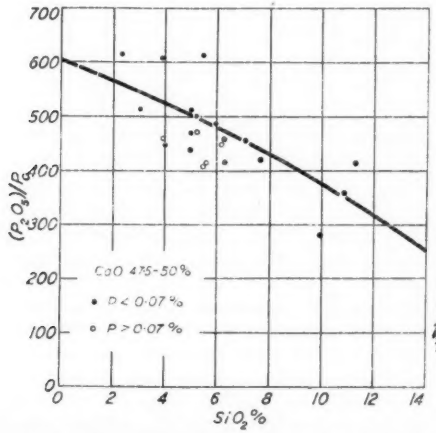


FIG. 17.—Distribution of phosphorus with slags of low silica content and containing 47·5 to 50 per cent. CaO.

The division is once more into groups of 42·5 to 45 per cent., 45 to 47·5 per cent., 47·5 to 50 per cent. and over 50 per cent. CaO. Here again the phosphorus content of the metal by reference to its manganese content has been recalculated in each case to the value it would have if the steel were free from manganese, and the ratio $(P_2O_5)/P_0$ plotted against the silica content.

The maximum range of scattering of the few individual values which for no apparent reason deviate from the mean is of the order of $\pm$ 15 per cent. and is probably to be attributed to the influence of temperature. If the temperature relations as found in Fig. 4 above for iron phosphate slags free from silica and lime were taken as a basis, the upper limit of the range of scattering would be about 1,610° and the lower about 1,575° C.

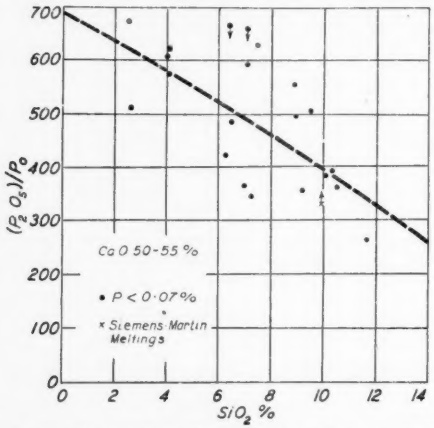


FIG. 18.—Distribution of phosphorus with slags of low silica content and containing 50 to 55 per cent. CaO.

In Figs. 15 to 17 the values are further classified according as the sample of steel in question contained less or more than 0·07 per cent. of phosphorus ; values obtained in excess of this amount lie only a very little lower than those corresponding to a smaller content of phosphorus and are still within the range of scattering. As the high values of phosphorus extended to 0·16 per cent. it may be said that, at least up to 0·10 per cent. phosphorus, the ratio V_{P_0} may for practical purposes be assumed as invariable even with a high lime content in the slag.

Fig. 19 is a composite diagram obtained by bringing together all the curves given in Figs. 8 to 18 ; with a view to clarity, however, the lime content instead of the silica content has been used for the abscissa. In this way the relation between $(P_2O_5)/P_0$,

here written as $(P_2O_5)/P_{0\,theor.}$ or $VP_{0\,theor.}$ and referring to the temperature of 1,600° C., and the lime content for each of a series of invariable silica contents is represented by a family of curves, enabling the effect on the distribution of phosphorus due

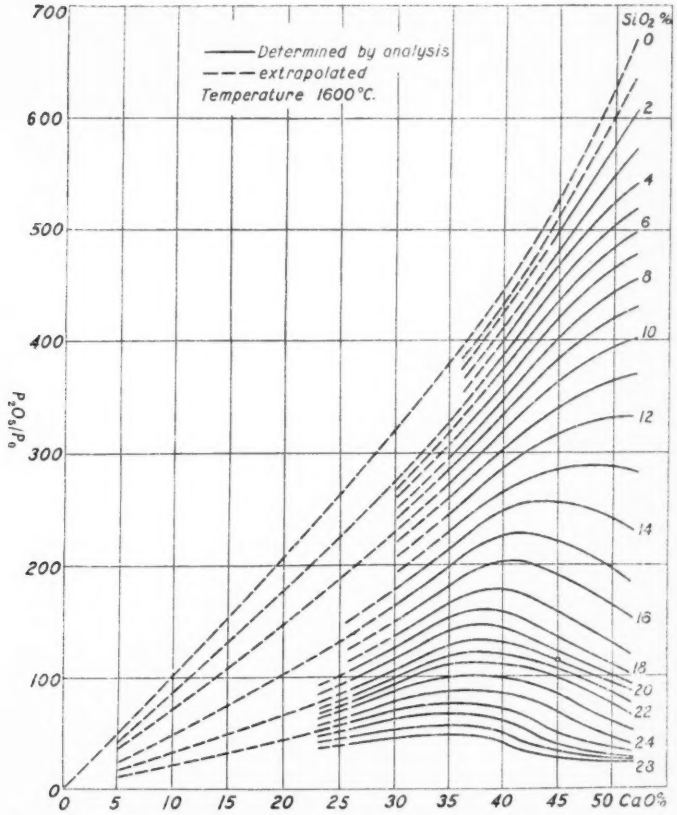


FIG. 19.—Composite diagram of the distribution of phosphorus between steel and slag in the basic process (after allowing for the manganese content of the steel by means of Fig. 7).

to any lime or silica present in the slag to be easily estimated either directly or by interpolation. Combining this with Fig. 7 it is possible to find the phosphorus content of the steel corresponding to any given combination of phosphoric acid, lime and

TABLE IV.—*Check Calculation of Composite Diagram and Comparison with the Most Accurate Determinations Hitherto Published.*

(1)	(2)	(3)	(4)	(5)	(6)	(7)	(8)	(9)	(10)	(11)	(12)	(13)	(14)	(15)	(16)	(17)	(18)
	Analyses of Steels.				Slag Analyses.												
No.	C. %	Mn. %	P. %	S. %	Fe. %	MnO. %	CaO. %	SiO_2. %	MgO. %	P_2O_5. %	Temperature. °C.	$\frac{(P_2O_5)}{P_e}$ theor. (1,600° C.*)	P_e	P Authors	P Schenck.	Δ Authors	Δ Schenck.
1	0·41	0·43	0·011	0·023	8·2	12·8	43·1	22·3	6·0	0·92	1585	103	0·0089	0·015	0·012	0·004	0·001
2	0·54	0·39	0·010	0·029	9·7	11·6	42·2	21·0	6·0	1·4	1582	117	0·0120	0·017	0·017	0·007	0·007
3	0·06	0·31	0·019	0·039	11·8	13·0	40·4	18·8	7·5	1·8	1627	146	0·0123	0·015	0·017	0·004	0·002
4	0·20	0·50	0·018	0·028	9·9	16·4	43·1	15·7	6·3	1·9	1615	208	0·0091	0·023	0·018	0·005	0·000
5	0·12	0·32	0·011	0·023	15·5	14·6	39·1	14·7	4·7	1·4	1602	234	0·0060	0·007	0·007	0·004	0·004
6	0·17	0·29	0·012	0·037	11·1	14·8	41·0	18·6	4·0	1·9	1594	146	0·0130	0·014	0·017	0·002	0·005
7	0·09	0·34	0·03	0·03	9·0	13·6	38·4	23·2	7·80	2·47	1657	96	0·0257	0·033	0·059	0·003	0·029
8	0·12	0·27	0·01	0·03	12·9	10·0	41·6	17·0	10·75	1·65	1621	175	0·0094	0·010	0·013	0·000	0·003
9	0·08	0·29	0·01	0·03	13·3	11·4	40·2	14·4	12·68	2·24	1657	246	0·0091	0·010	0·012	0·000	0·002
10	0·07	0·38	0·02	0·03	9·7	10·1	42·4	19·7	10·49	1·90	1646	128	0·0148	0·020	0·029	0·000	0·009
11	0·10	0·23	0·01	0·04	13·3	9·4	41·6	16·6	11·23	1·67	1657	190	0·0088	0·009	0·012	0·001	0·002
12	0·39	0·37	0·01	0·03	12·0	11·9	41·2	15·7	10·76	2·50	1641	213	0·0117	0·015	0·020	0·005	0·010
13	0·14	0·36	0·01	0·03	11·9	11·2	43·1	13·2	12·28	1·96	1631	278	0·0070	0·010	0·013	0·000	0·003
14	0·25	0·29	0·014	0·033	11·8	11·3	44·3	17·0	7·2	2·8	1600	163	0·0172	0·019	0·016	0·005	0·002
15	0·28	0·48	0·024	0·032	8·2	16·9	42·7	20·6	5·6	1·1	1600	119	0·0092	0·021	0·018	0·003	0·006
16	0·26	0·24	0·023	0·036	8·9	8·3	46·9	19·1	7·3	2·6	1575	113	0·0230	0·025	0·023	0·005	0·003
17	0·27	0·44	0·040	0·034	7·4	11·8	44·9	19·1	6·4	3·3	1585	122	0·0270	0·046	0·048	0·006	0·008
18	0·41	0·24	0·019	0·032	11·8	7·5	45·7	17·4	7·2	3·0	1595	145	0·0207	0·022	0·015	0·003	0·004
19	0·32	0·25	0·012	0·028	13·2	7·3	48·1	14·5	5·9	2·1	1595	227	0·0092	0·010	0·008	0·002	0·004
20	0·65	0·38	0·015	0·021	10·5	10·1	45·1	15·7	10·7	1·8	1590	201	0·0089	0·013	0·011	0·002	0·004
21	0·69	0·38	0·010	0·025	12·4	14·9	43·8	13·2	6·0	2·1	1580	280	0·0075	0·011	0·009	0·001	0·001
22	0·56	0·31	0·010	0·022	15·0	10·4	46·3	10·8	7·3	1·7	1575	354	0·0048	0·006	0·009	0·004	0·001
23	..	0·69	0·045	0·029	7·60	10·7	43·52	20·8	..	1·09	..	115	0·0095	0·060	0·027	0·015	0·018
24	..	0·44	0·030	0·055	8·53	14·0	41·44	20·2	..	1·47	..	126	0·0117	0·021	0·031	0·009	0·001
25	..	0·58	0·052	0·053	10·11	20·0	36·6	21·0	..	0·95	..	133	0·0071	0·035	0·023	0·017	0·029
26	..	0·42	0·018	0·045	8·03	10·8	43·3	19·2	..	0·95	..	128	0·0074	0·013	0·019	0·005	0·001
27	..	0·51	0·038	0·041	8·76	12·8	39·9	21·6	..	0·95	..	114	0·0083	0·025	0·022	0·013	0·016

* Mean Temperature.

TABLE IV.—(*continued.*)

(1)	(2)	(3)	(4)	(5)	(6)	(7)	(8)	(9)	(10)	(11)	(12)	(13)	(14)	(15)	(16)	(17)	(18)
No.	Analyses of Steels.				Slag Analyses.						Temperature. °C.	$\frac{(P_2O_5)}{P_o}$ theor. (1600 C.°)	P_o	P Authors	P Schenck.	Δ Authors	Δ Schenck.
	C. %	Mn. %	P. %	S. %	Fe. %	MnO. %	CaO. %	SiO_2. %	MgO. %	P_2O_5. %							
28	..	0·20	0·043	0·048	11·2	5·7	50·1	8·9	..	17·5	..	429	0·0408	0·043	0·038	0·000	0·005
29	..	0·25	0·033	0·047	9·8	4·9	51·1	8·6	..	18·3	..	440	0·0416	0·045	0·048	0·012	0·015
30	..	0·20	0·051	0·047	11·8	6·3	45·7	7·4	..	19·8	..	424	0·0467	0·049	0·060	0·002	0·009
31	..	0·16	0·042	0·055	12·4	5·8	46·7	8·1	..	17·9	..	420	0·0426	0·044	0·040	0·002	0·002
32	..	0·23	0·061	0·050	8·7	6·0	48·3	9·3	..	20·8	..	406	0·0512	0·055	0·110	0·006	0·049
33	..	0·17	0·054	0·058	10·2	6·3	46·0	9·8	..	19·0	..	380	0·0500	0·052	0·084	0·002	0·030
34	..	0·25	0·049	0·036	9·6	6·9	50·3	6·1	..	20·9	..	486	0·0430	0·047	0·059	0·002	0·010
35	..	0·26	0·049	0·040	9·7	6·7	50·3	7·4	..	19·9	..	460	0·0433	0·048	0·057	0·001	0·008
36	..	0·42	0·06	0·035	9·00	5·33	53·5	4·09	1·83	22·4	..	550	0·0407	0·070	0·046	0·010	0·014
37	..	0·37	0·051	0·036	10·26	5·45	50·2	4·00	1·39	22·4	..	528	0·0425	0·061	0·047	0·010	0·004
38	..	0·36	0·052	0·037	10·10	5·38	49·0	3·85	1·01	23·0	..	520	0·0442	0·062	0·065	0·010	0·013
39	..	0·31	0·047	0·032	9·82	5·51	52·0	4·18	2·31	21·8	..	540	0·0404	0·048	0·056	0·001	0·009
40	..	0·34	0·057	0·038	9·36	6·76	55·1	2·60	1·53	21·9	..	620	0·0353	0·043	0·029	0·014	0·028
41	..	0·40	0·058	0·033	11·32	7·96	46·8	2·40	1·09	24·9	..	520	0·0480	0·081	0·088	0·023	0·030
42	..	0·45	0·072	0·036	9·70	7·58	41·9	2·51	1·14	24·9	..	590	0·0422	0·082	0·054	0·010	0·018
43	..	0·40	0·06	0·035	9·70	6·88	49·9	2·31	1·25	23·3	..	570	0·0409	0·065	0·060	0·000	0·000
44	..	0·10	0·068	0·048	13·47	5·4	43·1	5·52	..	21·25	..	425	0·0500	0·051	0·042	0·017	0·026
45	..	0·03	0·057	0·058	19·4	4·8	35·7	5·21	..	20·05	..	333	0·0604	0·060	0·038	0·003	0·019
46	..	0·24	0·070	0·047	10·9	5·6	44·9	5·07	..	25·0	..	454	0·0551	0·062	0·066	0·008	0·004
47	..	0·24	0·054	0·039	11·7	5·2	45·4	4·46	..	24·8	..	470	0·0528	0·058	0·033	0·004	0·019
48	..	0·48	0·031	0·073	13·4	9·4	44·5	11·4	..	7·84	..	328	0·0239	0·046	0·027	0·015	0·004
49	..	0·52	0·025	0·060	15·8	8·1	49·8	10·0	..	6·50	..	395	0·0164	0·041	0·016	0·016	0·009
50	..	0·47	0·020	0·077	15·6	6·5	46·5	10·4	..	6·00	..	365	0·0164	0·032	0·017	0·012	0·003
51	..	0·47	0·040	0·067	13·2	10·3	46·3	12·4	..	5·00	..	310	0·0161	0·032	0·018	0·008	0·022
52	..	0·14	0·017	0·044	10·93	5·7	49·34	10·6	..	15·66	..	372	0·0421	0·044	0·047	0·027	0·030
53	..	0·12	0·030	0·050	11·07	6·0	49·54	10·1	..	15·38	..	390	0·0394	0·040	0·044	0·010	0·014
54	..	0·12	0·030	0·050	13·42	5·8	49·74	10·2	..	15·16	..	387	0·0392	0·040	0·024	0·010	0·006

* Mean Temperature.

silica in the slag and of manganese in the steel; for instance, if the slag contains 45 per cent. of CaO and 20 per cent. of SiO_2 it is seen in Fig. 19 (as marked by a circle) that $(P_2O_5)/P_0$ has the value 114, and if, for example, there is 3 per cent. of P_2O_5 the amount of P_0 will be $3/114 = 0{\cdot}026$ per cent. If, further, the steel happens to contain 0·40 per cent. of manganese, the ordinate 0·026 is to be sought in Fig. 7 and followed along the curve as far as the point corresponding to the named amount of manganese, where the actual phosphorus content of 0·038 per cent. may be read off. In a similar way P_2O_5 can be found from P with known quantities of manganese, silica and lime.

Calculations of the phosphorus content based on slag analyses and on the manganese content have been made, using the method described above, from a collection of analyses by H. Schenck[6] (*loc. cit.* pp. 156–157) which were not used in compiling Fig. 19, and the results obtained are given in column 15 of Table IV., whilst column 16 contains the phosphorus contents as calculated by Schenck himself. The method of calculation used by the latter, based on extremely complicated thermodynamical formulæ, was the one which had hitherto come closest to actual conditions. A comparison of the values obtained by both methods of calculation with those found from chemical analysis shows the following deviations (Δ authors and Δ Schenck. in columns 17 and 18 respectively) :

	Mean deviation.	Maximum deviation.
Maurer and Bischof's method .	0·007 per cent.	0·027 per cent.
Schenck's method . . .	0·011 „	0·049 „

It is indisputable from these figures that the present graphical method is of very wide application and is preferable to purely numerical methods such as that of Schenck. The former also has the advantage of being immediately usable by the practical man in charge.

(c) *Temperature Conditions Affecting the Phosphorus Distribution in the Basic Process of Steel Manufacture.*—In the comparisons given above the influence of temperature has been ignored. A temperature effect, however, is present not only with pure iron phosphate slags but also with those containing lime or silica, and this is made more particularly apparent by the fact that the $(P_2O_5)/P_0$ ratio rises steeply for a high carbon content in the steel. To examine this matter all the analyses of steel and slag included in the collection by Maurer and Bischof,[2] which has already been mentioned several times, were reviewed in the light of temperature data, those values of $(P_2O_5)/P_0$ which had been ascertained on the basis of the manganese and phosphorus contents of the steel and the phosphorus content of the slag

being divided by those obtained only on the theoretical basis of the lime and silica contents in Fig. 19. The respective ratios were then plotted against the temperatures in Fig. 20. Despite a fairly pronounced scattering of the points it is possible to perceive a consistent drop in the ratio V_{P_0} and its theoretical equivalent $V_{P_0\,\text{theor.}}$ as the temperature rises; in other words, for equal phosphorus contents in two slags the phosphorus content in the steel is greater in the case of the slag which has the higher temperature—a fact already known qualitatively from practical experience of operating. This effect of temperature on the phosphorus distribution does not seem to depend much on the lime and silica contents of the slag, so far as can be inferred from Fig. 20 for silica contents of 15 to 25 per cent. and lime contents of 35 to 45 per cent. This is confirmed by the following considerations:

Referring to Fig. 4, which gives the temperature effect on the phosphorus distribution in the presence of pure iron phosphate slag, at 1,600° C. the value of $(P_2O_5)/P_0$ is approximately 2 and at 1,550° approximately 3·6; hence the ratio V_{P_0}: $V_{P_0\,\text{theor.}} = 3{\cdot}6 : 2 = 1{\cdot}8$. Now plotting 1 at 1,600° and 1·8 at 1,550° we obtain the broken line in Fig. 20 which gives the temperature relationship of the phosphorus distribution for pure iron phosphate slag, and there can be no doubt as to the close agreement of this with results obtained from actual slags. If, then, temperature is to be taken into account, the curve in Fig. 20 will (at any rate provisionally) provide the method of doing so which most nearly corresponds with all the conditions obtaining in the basic open-

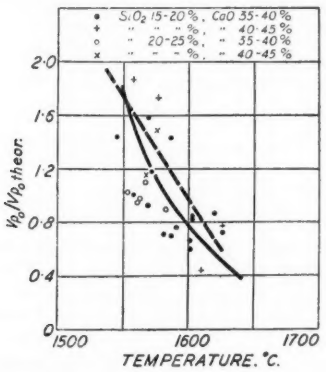


FIG. 20.—Relation of phosphorus distribution to temperature in the presence of basic Siemens-Martin slags compared with pure iron phosphate slags (broken line, from Fig. 4).

hearth process. For example, to ascertain the ratio of distribution at any given temperature with a given lime and silica content the $(P_2O_5)/P_0$ ratio is taken from Fig. 19 and is multiplied by the quotient appropriate to the temperature in question, as given in Fig. 20.

Even for the slags poor in silica that occur in the Thomas process the assumption of 1,600° C. for the final temperature should be admissible, at least to a considerable extent, seeing that the temperatures here met with are sometimes put at 1,575° C.[13] and sometimes at 1,650° or more.[14] The temperature relationship given in Fig. 20 should therefore hold good for the Thomas process also.

(*d*) *Effect of Manganous Oxide in the Slag on the Distribution of Phosphorus.*—In order to establish the effect of manganous oxide a procedure was followed similar to that used in checking the temperature relationship : The ratios $(P_2O_5)/P_0$ were determined first on the basis of the manganese and phosphorus in the steel and of the phosphorus pentoxide in the slag (VP_0), then on the theoretical basis of the lime and silica in the slag alone according to Fig. 19 ($VP_{0\,theor.}$) ; the first result was divided by the second, and the quotient was plotted against the amount of manganous oxide present. Fig. 21 shows an

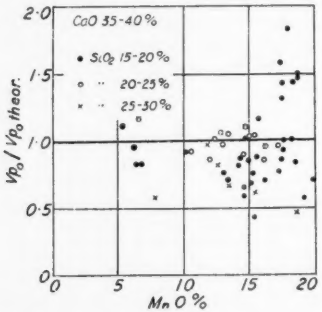
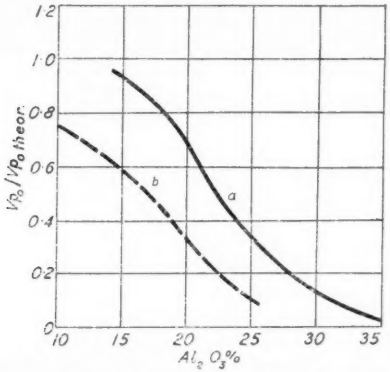


Fig. 21.—Effect of manganous oxide on the phosphorus distribution.

example of this kind of check, having reference to slags containing 35 to 40 per cent. of CaO. Here, as in the case of greater and smaller lime contents, no law of relationship could be determined—a result contrary to what was found true of the influence of the manganese content of the steel itself. It follows that neither the opinion of Schenck[6] (*loc. cit.* p. 159) that dephosphorisation is promoted by the presence of manganous oxide nor that of Schackmann and Krings[4] (*loc. cit.* p. 176) to the effect that reduction of phosphorus is aided by this agency is tenable. Probably

manganous oxide behaves towards phosphorus pentoxide exactly as does ferrous oxide in the matter of the formation of phosphates and in relation to the oxidation of the phosphorus in the steel.

(*e*) *Effect of Magnesia in the Slag on the Distribution of Phosphorus.*—The effect of magnesia was studied in the same way as that of manganous oxide. It has already been inferred from Figs. 8 to 14 that magnesia contents of up to 10 per cent. cause no change in the ratio $(P_2O_5)/P_0$, and the checking of a large number of melts containing from 10 to 15 per cent. of magnesia showed no noticeable change even with these higher proportions. These statements, however, apply only for the range of slag formation that occurs in the basic open-hearth process, and it was not possible to find out the effect of magnesia in the low-silica slags of the Thomas process. Schackmann and Krings[4] found that the addition of magnesia to pure iron phosphate slags aided dephosphorisation in the same direction as lime.



(*a*) Curve derived from analyses of melts in basic electric and Siemens-Martin furnaces.
(*b*) Comparative curve from Fig. 2B.

FIG. 22.—Effect of alumina on the phosphorus distribution.

(*f*) *Effect of Alumina on the Distribution of Phosphorus.*—According to the experimental melts made by Schackmann and Krings the effect of alumina is to cause a marked change in the ratio $(P_2O_5)/P_0$ in the same direction as silica, and the same unfavourable effect of alumina on dephosphorisation has been observed in the electric arc furnace and in the Siemens-Martin furnace. The present authors' earlier work[2] included a series of steel and slag samples from these two types of furnace having high alumina contents. Now the proportions found in these samples, first by analysis and secondly from Fig. 19, give very

low quotients when divided by one another ; in other words, the higher the alumina content in a slag of otherwise unaltered composition the greater will be the amount of phosphorus found in the steel. The low values of the quotients obtained from the electric process of steel manufacture cannot be attributed to peculiar conditions of temperature because the same observation is true of the Siemens-Martin process.

The quotients are plotted against the alumina content in Fig. 22 for comparison, the samples from the electric arc furnace giving results approximately in accordance with the full-line curve derived by adjustment of the $(P_2O_5)/P_0$ ratio, taking account of the alumina. In order to allow comparison with the effect of alumina in pure iron phosphate slags the curve shown in Fig. 2B has been re-calculated accordingly and has then been embodied in Fig. 22 as a broken line.

The tendency of alumina (considered in relation to the amount present) is therefore to cause the same alteration of the ratio $(P_2O_5)/P_0$ for slags such as occur in steel manufacture as was found in the case of the experimental melts with pure iron phosphate slags.

IV.—Concluding Remarks.

It is pertinent to enquire what aim is being pursued in the results that have been obtained, and in the work that has been done to date, in studying the reactions attending the manufacture of steel.

Contributions to the literature of the subject[15] are already available which throw light on the influence of slag formation and slag composition upon the quality of the resulting steel. But the mechanism of that influence is not yet understood, nor is it known how far the oxides dissolved in the iron alter in composition and quantity according to the nature of the slag. In steelmaking, the aim must always be to secure slags of such a kind as will, so far as is possible, produce steel of the highest quality ; but deliberate control over slag formation must be preceded by knowledge of the principles on which it depends.

This is an aim which, undoubtedly, has not yet been attained ; yet the foregoing contribution may be of some service in clearing the way to its attainment.

V.—Summary.

By means of melting tests, the laws of the distribution of phosphorus were established, first in outline and then in relation to the presence of pure iron phosphate slags, confirming the authors' earlier conclusion that where the phosphorus contents of the iron and of the slag are low the distribution at any given tempera-

ture is invariable. Further, a certain temperature relationship was established.

Additions of silica to the slag were found to cause a reduction of the ratio of distribution $V_P = (P_2O_5)/P$.

Next, in order to study dephosphorisation in the process of steel manufacture, the effect of changing proportions of manganese in the steel was determined by curves and was shown to be important, an increase in the manganese while the composition of the slag remained constant resulting in an increase in the phosphorus content of the steel.

The lime content being altered only by small steps, the ratio of distribution V_{P_0} as governed by the amount of silica was determined by convergence. It was found that the values of V_{P_0} increased with increasing amounts of lime, and that the effect of the latter was diminished by the silica. A fundamental diagram was compiled enabling the phosphorus content of the steel to be theoretically determined for any given composition of slag and procedure in steel making.

The temperature relationships of the slags occurring in practice were found to correspond completely with those established for pure iron phosphate slags. As the temperature rises dephosphorisation is rendered less effective—a fact already recognised qualitatively.

Within the limits of the slags met with in operation, manganous oxide and magnesia are without effect on the distribution of phosphorus. Alumina has an unfavourable effect on the ratio of distribution, which is exercised in the same direction as is the case for pure iron phosphate slags.

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THE PENETRATION OF MOLTEN WHITE METALS INTO STRESSED STEELS.*

By W. E. GOODRICH, M.Met. (Sheffield).

Synopsis.

An investigation was carried out to determine the resistance to interpenetration of various types of steel, subjected to stress in the presence of molten white metals, with reference to temperature variations, rate of loading, chemical composition, tensile properties and microstructure. It was found that premature failure only occurred when the samples were stressed in tension in the presence of molten white metals, and that resistance to penetration decreased with rise in temperature and with slower rates of loading.

Consideration of the test results on carbon, nickel, nickel-chromium, nickel-chromium-molybdenum and austenitic steels, tested at 250° C. and 350° C., led to the conclusion that the static tensile properties were no absolute indication of the interpenetration-resistant properties. The addition of nickel up to 1·2 per cent., at least, to carbon steels was deleterious at both temperatures, whilst the addition of molybdenum to nickel-chromium steels appeared to confer no marked advantage.

The microstructure of the material constituted the main factor influencing the resistance to interpenetration. With carbon and nickel steels, the presence of free ferrite was a distinct advantage at 250° C., particularly when in the form of a network, but at 350° C. increased amounts of free ferrite were of little benefit. With reference to nickel-chromium and nickel-chromium-molybdenum steels, the presence of intermingled ferrite was again advantageous at the lower temperature. At 350° C., however, there appeared to be an optimum amount of intermingled ferrite which was of advantage, the presence of ferrite in excess of that optimum quantity reducing the resistance of the alloy steels to interpenetration. One of the nickel-chromium steels, with a microstructure consisting of small-grained compact sorbite and a small amount of intermingled ferrite, showed remarkably good penetration-resistance at both temperatures.

The effect of grain-size was very marked, small grain-size being beneficial, particularly at 250° C.

Temper-brittleness appeared to have no effect on the resistance to interpenetration.

Introduction.

Cases have been recorded in which localised failure of various classes of steel had taken place under stresses considerably lower than the maximum stress of the material as determined under normal testing conditions. Microexamination indicated the penetration of foreign molten metal, the presence of which was attributed to soldering or brazing operations or seized bearings.

With such premature failures in mind, and in order to prevent or minimise their re-occurrence, it was considered desirable to

* Received May 29, 1935.

investigate the effect of such variables as temperature, rate of application of load, white metals of differing chemical composition, tensile properties, and degree of temper-brittleness in the case of steels of widely varying analysis and microstructure, and an attempt has been made in this paper to draw some conclusions regarding the effect of those variables on the resistance of steels to the penetration of molten white metals.

Description of Apparatus and Preliminary Experiments.

For the purpose of these experiments, rectangular pieces of the various steels were prepared 3·25 in. long × 0·4 in. × 0·3 in. During the testing operations, each sample was supported, with the 0·3 in. face horizontal, between knife-edges 2·7 in. apart, a specially prepared ram terminating in a knife-edge being fitted into a hydraulic Brinell hardness testing machine, in place of the usual 10-mm. ball and holder, in such a manner that the applied load acted on the test-piece at a position midway between the bottom knife-edges. The test-piece under loading test was heated to the required temperature by two small electric furnaces supported by the piece itself and placed one on each side of the centre ram. The furnaces consisted of windings of nichrome wire on small nickel sheet formers, with interposed asbestos sheet, the furnaces being lagged as much as permissible with asbestos cord impregnated with alundum cement. The test-pieces were all maintained at constant temperature during the testing operations, and the deflection of the piece for each applied load was measured, utilising an optical lever system with a magnification of 75 times. A photograph of the apparatus is shown in Fig. 12 (Plate I.). Samples of the various steels were tested at 250° C. and 350° C. without molten white metal present ; at 250° C. in the presence of a lead-tin solder ; and at 350° C. in the presence of molten bearing metal. The analyses and thermal data of the white metals were as follows :

	Solder.	Bearing Metal.
Lead	37·0%	..
Tin	63·0%	88·6%
Antimony	..	7·75%
Copper	..	3·58%
Eutectic melting point	182·5° C.	238·0° C.
Upper freezing point	182·5° C.	324·0° C.

In order to determine whether the flux used had any effect on the penetration of the white metals, and to decide on the effects of compressive and tensile stresses, experiments were carried out on samples of steel under the following testing conditions :

(1) Without solder, bearing metal or flux.
(2) With " Coraline " soldering paste but no white metal.
(3) With flux and an appreciable amount of lead-tin solder on the *upper* surface ; temperature 250° C.

(4) With flux and lead-tin solder on the *under* surface; temperature 250° C.
(5) With flux and bearing metal on the *upper* surface; temperature 350° C.
(6) With flux and bearing metal on the *under* surface; temperature 350° C.

From those tests it was concluded that, up to the maximum load withstood by the specimens (defined as that load at which the deflection of the samples was so pronounced that even rapid pumping failed to effect a load increase), all samples without molten metal in contact with the face under tension behaved in a similar manner and remained unbroken, and that the presence of flux or molten metal on the surface subjected to compressive stresses had no effect on the strength of the sections under test. The presence of molten white metal in conjunction with *stresses in tension*, however, in every case caused fracture at loads appreciably less than those applied to the uncoated samples. Later experiments indicated that the effect of molten solder was exactly similar to the effect of molten bearing metal at 350° C.

Referring to Fig. 15 (Plate II.), sample 1 shows the amount of deflection applied to a 1 per cent. nickel steel (Mark *DI*), unsoldered, and tested at 250° C., no sign of cracking being discernible. Samples 2 to 6 inclusive, show the fractures typical of the same steel tested under similar conditions but with the application of molten solder to the surface subjected to tensile stress. Sample 7 is a photograph of a test-piece of low-carbon steel, showing that much greater deflection was obtained before fracture occurred. The molten metal is clearly discernible, in most cases, to the extremity of the fracture.

Figs. 16 and 17 (Plate III.) show the typical structures of an oil-hardened and tempered carbon steel (*CA*) and a nickel-chromium-molybdenum steel (*VX*), respectively, when etched in alcoholic nitric acid. From Fig. 16 it is apparent that in the case of the carbon steel, the molten white metal traversed paths at the junction of ferritic and sorbitic areas, although there are isolated instances where the molten metal had crossed sorbitic areas. Fig. 17 indicates quite clearly that the molten metal followed the outline of the polygonal grain boundaries. These examples are typical illustrations of the manner of penetration in steels of microstructures similar to *CA* and *VX*, respectively, whether tested under stress at 250° C. or 350° C.

Experiments were necessary, before commencing the investigation on various steels, to determine what effects were to be expected with different rates of application of stress.

Four different types of steels were chosen, *viz.*:

CA, a typical oil-hardened and tempered carbon axle steel.
DI, a typical oil-hardened and tempered 1 per cent. nickel steel.
TR, an oil-hardened and tempered nickel-chromium-molybdenum steel, composed of extremely large grains of acicular sorbite with intermingled ferrite.
VX, an oil-hardened and tempered nickel-chromium-molybdenum steel, composed of small grains of loose sorbite and intermingled ferrite.

Samples prepared from these steels were tested at 250° C., both with and without molten solder present, using a variety of loading rates, the load being increased by 50-kg. increments, and each increment maintained for the desired time interval. The rate of loading for each 50-kg. increase was kept constant throughout all individual tests. It was found that under the lower loads, for any one test-piece, the deflection at the end of the specified time

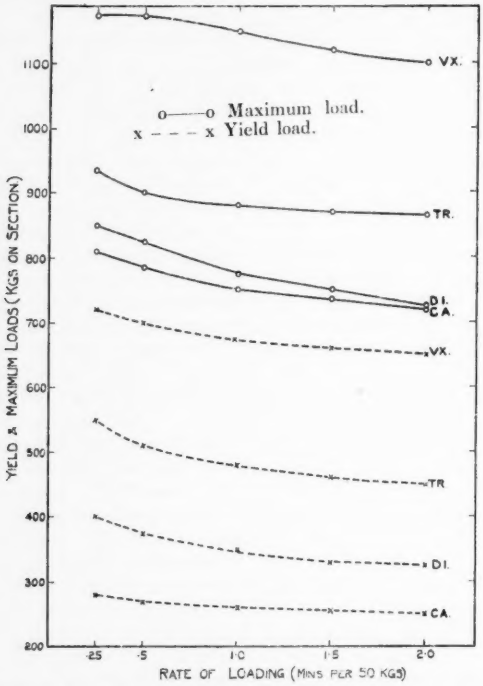

FIG. 1.—Effect of Rate of Loading on Yield and Maximum Loads. Tested at 250° C. (solder present).

interval was the same as that at the commencement of that interval. In every case, however, a load was eventually reached at which the deflection at the end of the interval was slightly in excess of the deflection noted at the first application of that load ; in other words, the test-piece had " sagged " slightly under constant load during that interval. The load at which this slight sagging was first discernible is referred to throughout this paper

as the "yield load," and, when calculated according to the usual formula for bending stresses under this system, the yield load was found to be almost identical with the yield point obtained in the ordinary static tensile test. The effect of variations in the rate of loading on the yield load, maximum load and deflection at fracture on samples tested at 250° C. is shown graphically in Figs. 1 and 3. From Fig. 1 it is evident that there was a gradual

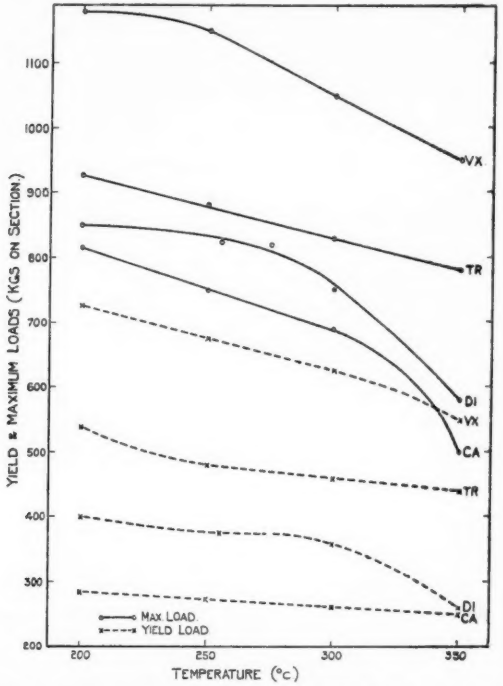

FIG. 2.—Effect of Temperature on Yield and Maximum Loads. Rate of Loading 50 kg. per min. (solder present).

decrease in the yield and maximum loads in all cases, the decrease being more pronounced for rates of 1 min. or less per 50-kg. load increments and then more gradual as the time for each load increment was increased.

Soldered test-pieces from each of the three steels *DI*, *TR* and *VX* showed that with a rate of loading of 2 min. per 50-kg. increment the pieces broke at loads of 725, 865 and 1,100 kg.,

respectively. Samples of the three steels were then loaded up to loads considerably less than the figures indicated above, with the following results :

Sample.	Testing Temp.	Load.	Broke in—
DI	250° C.	675 kg.	10 hr.
TR	250° C.	800 ,,	7·5 ,,
VX	250° C.	1000 ,,	6·5 ,,

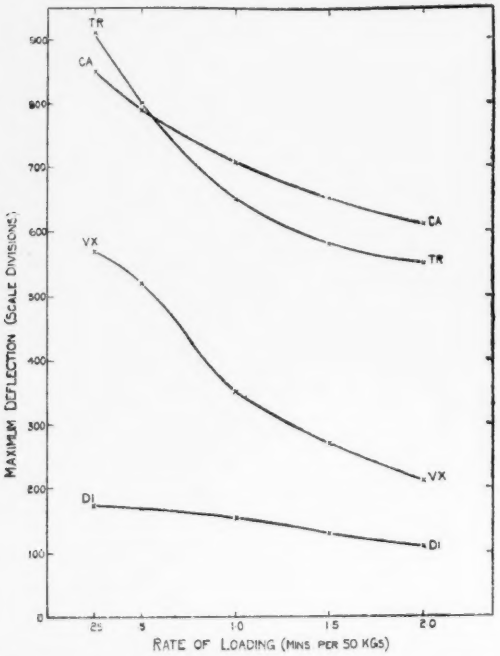

FIG. 3.—Effect of Rate of Loading on Maximum Deflection before Fracture. Tested at 250° C. (solder present).

It is evident from these tests that figures obtained on subsequent tests, under fixed conditions of temperature and rate of loading, should only be taken as comparative and not as definite indications of the stresses which could be applied with safety to any particular steel in service.

A much more important effect due to variations in rate of loading, however, is shown in Fig. 3. In this graph the maximum deflection at the time of fracture has been plotted against the rate of loading, and the comparatively small deflection of the 1 per

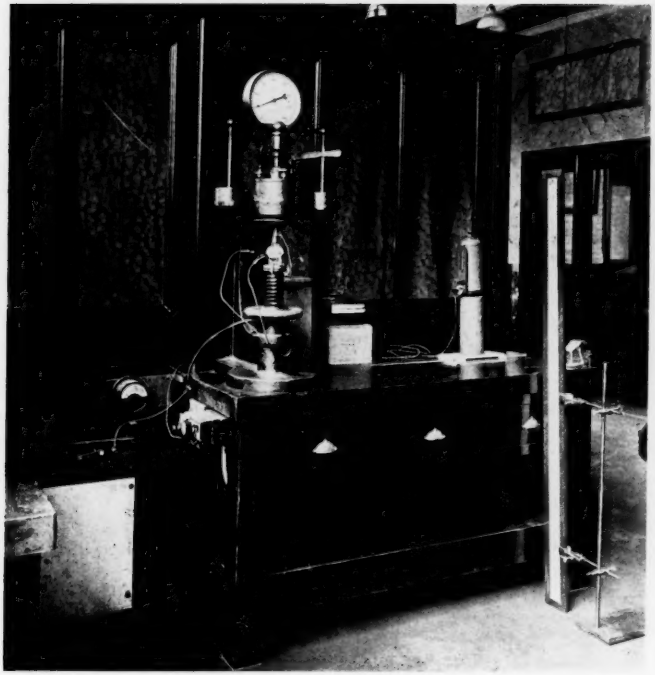

FIG. 12.—The Apparatus.

FIG. 13.—Austenitic Steel *W*. Very coarse grains of austenite. × 250.

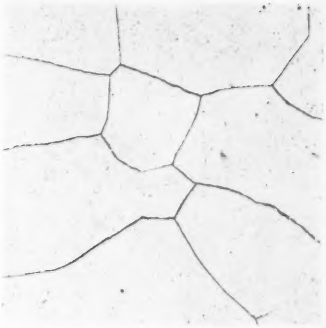

FIG. 14.—Austenitic Steel *WA*. Much smaller austenitic grains than steel *W*. × 250.

(Micrographs reduced two-thirds linear in reproduction).

[*Goodrich.*

[*To face p.* 48.

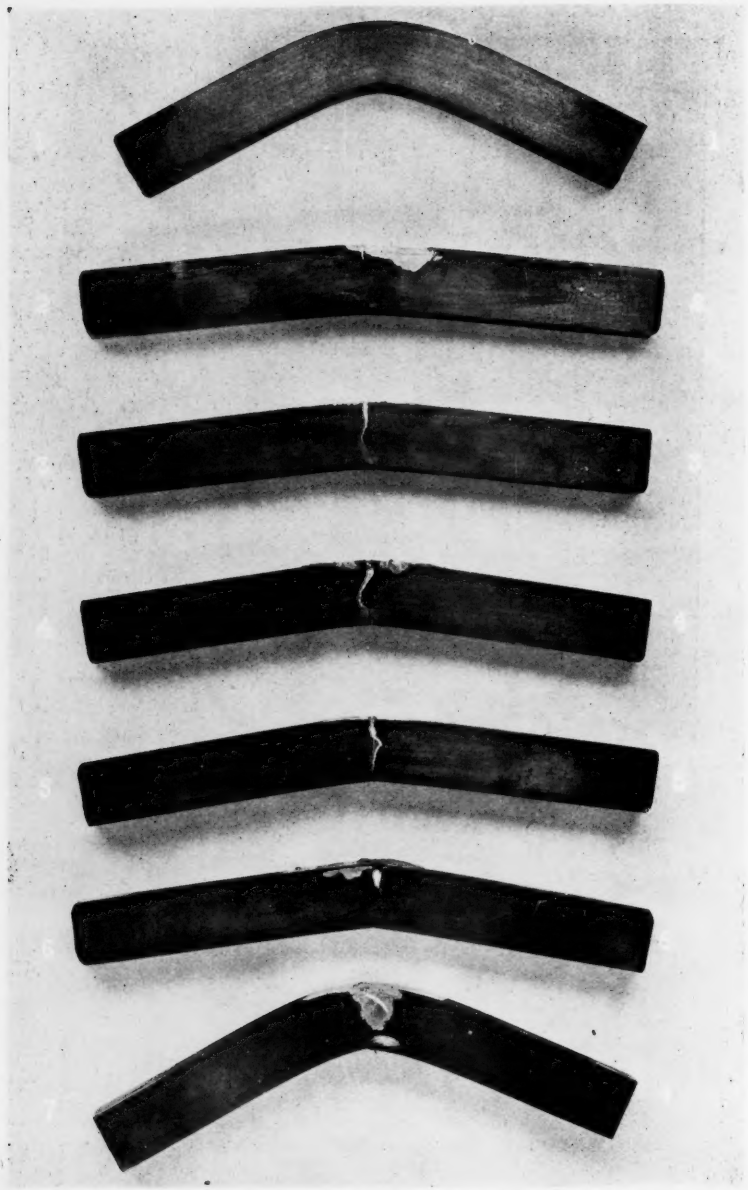

FIG. 15.—Photographs of Soldered Specimens, showing typical fractures. *Goodrich.*

FIG. 16.—Carbon Steel *CA*. Penetration of molten metal at junctions of ferritic and sorbitic areas. × 250.

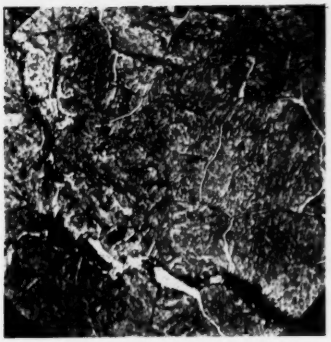

FIG. 17.—Ni-Cr-Mo Steel *VX*. Penetration of molten metal round the grain boundaries. × 250.

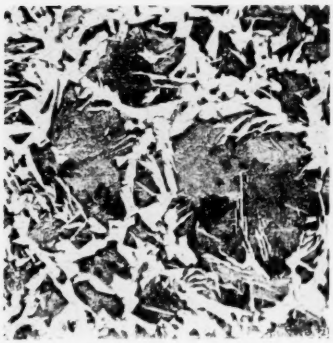

FIG. 18.—0·43% Ni Steel *A*. Polygonal grains of sorbite with ferritic boundaries. × 250.

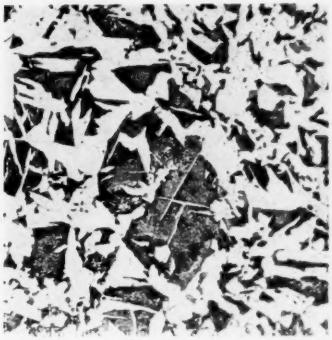

FIG. 19.—0·43% Ni Steel *AI*. Irregular and more massive formation of the ferrite than steel *A*. × 250.

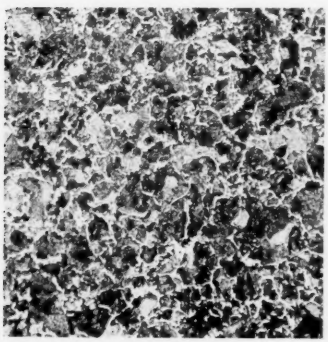

FIG. 20.—1·2% Ni Steel *D*. Ferrite as network round sorbitic grains. Less free ferrite than steel *A*. × 250.

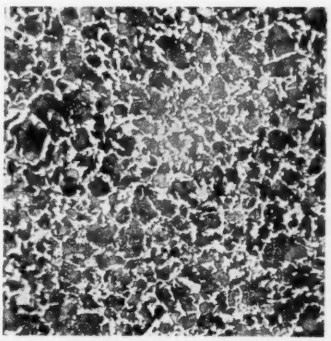

FIG. 21.—1·2% Ni Steel *DI*. Irregular and more massive ferrite than steel *D*. Less free ferrite than stee *AI*. × 250.

(Micrographs reduced to two-thirds linear in reproduction.) [*Goodrich*.

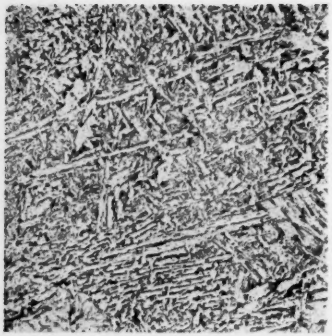

FIG. 22.—Ni–Cr Steel *G*. Coarse grains of loose sorbite and intermingled ferrite. × 250.

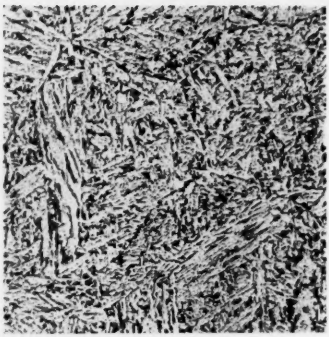

FIG. 23.—Ni–Cr Steel *H*. Structure similar to steel *G*, but larger grain-size. × 250.

FIG. 24.—Ni–Cr Steel *J*. Small grains of loose sorbite and intermingled ferrite. × 250.

FIG. 25.—Ni–Cr Steel *K*. Coarse grains of fairly loose sorbite and intermingled ferrite. × 250.

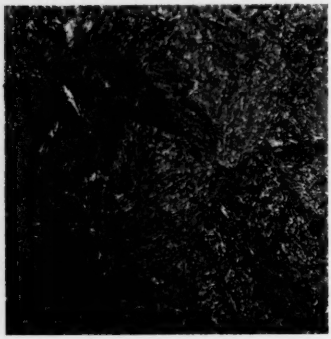

FIG. 26.—Ni–Cr Steel *L*. Coarse grains of sorbite. Less intermingled ferrite than steel *K*. × 250.

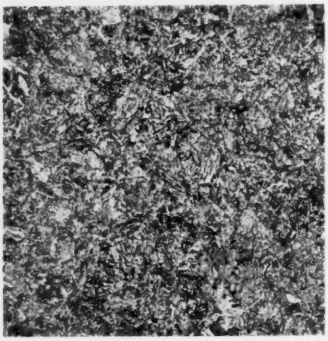

FIG. 27.—Ni–Cr Steel *M*. Fine-grained structure of compact sorbite. Smaller amount of intermingled ferrite than steel *J*. × 250.

(Micrographs reduced to two-thirds linear in reproduction.) [*Goodrich*.

FIG. 28.—Ni–Cr–Mo Steel *SB*. Coarse grains of loose sorbite and intermingled ferrite. × 250.

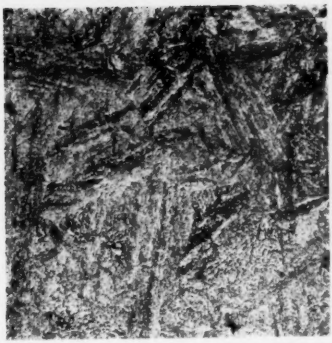

FIG. 29.—Ni–Cr–Mo Steel *SM*. Coarse grains of sorbite with less intermingled ferrite than steel *SB*. × 250.

FIG. 30.—Ni–Cr–Mo Steel *UB*. Small sorbitic grains with intermingled ferrite. × 250.

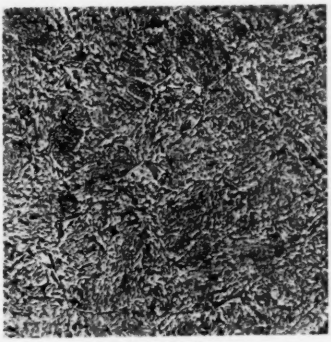

FIG. 31.—Ni–Cr–Mo Steel *UM*. Small grains of sorbite with more intermingled ferrite than steel *UB*. × 250.

FIG. 32.—Ni–Cr–Mo Steel *RB*. Annealed structure, consisting of loose sorbite and massive free ferrite. × 250.

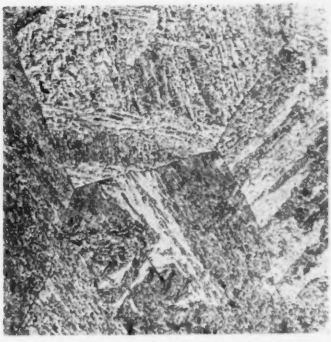

FIG. 33.—Ni–Cr–Mo Steel *TR*. Very coarse sorbitic grains with intermingled ferrite. Very pronounced grain boundaries. × 250.

(Micrographs reduced to two-thirds linear in reproduction.) [*Goodrich.*

cent. nickel steel (*DI*) is very apparent. The evidence shows, first, that the more quickly the load increments were applied the greater was the maximum permissible deflection ; and, secondly, that the decrease in permissible deflection without fracture was more rapid in the case of the alloy steels *TR* and *VX* than with the carbon steel *CA*. The amount of deflection permissible before fracture occurs is of considerable importance, since in

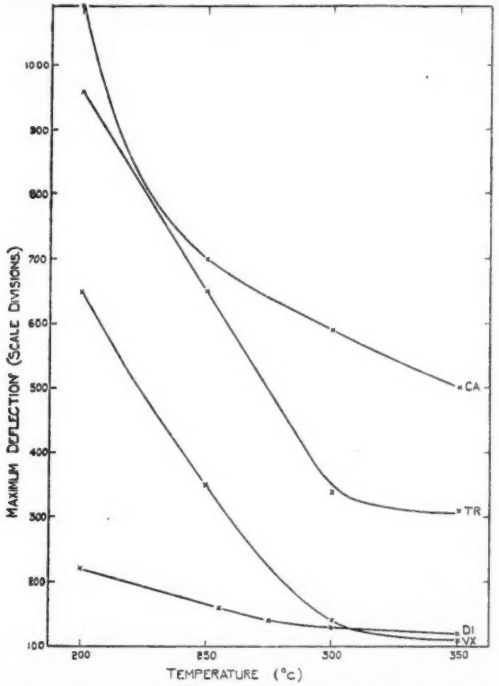

FIG. 4.—Effect of Temperature on Maximum Deflection before Fracture. Rate of loading, 50 kg. per min. (solder present).

steels of corresponding maximum stress, the greater that deflection the greater is the area enclosed by the load-strain curve and hence the energy required to cause fracture. Also, in those cases where the material can flow plastically to an appreciable extent, there is less possibility of localised stresses increasing to such a magnitude that failure due to penetration might occur.

A preliminary investigation was also conducted to determine the effect of the testing temperature with constant rate of loading,

on soldered samples, the results being shown graphically in Figs. 2 and 4. From Fig. 2 it is apparent that the yield and maximum loads of the steels decreased as the temperature of testing increased, whilst Fig. 4 shows the correlation of temperature and deflection in the four steels, the very marked effect of temperature change being noticeable.

As a result of the evidence brought out by these preliminary tests, it was decided that the following general testing procedure should be adhered to for comparative tests on the various types and conditions of steel under investigation :

(*a*) Temperatures of testing : (i) 250° C. with and without solder.
(ii) 350° C. with and without bearing metal.
(*b*) Rate of Loading : 1 min. per 50-kg. increment.
(*c*) Yield load, maximum load, maximum deflection and time to fracture to be measured.
(*d*) Load-deflection curves to be plotted to a common scale, and the area enclosed by the curves to be measured by a planimeter. These areas are recorded in the tables under the heading of "Curve Area."

All subsequent tests were carried out in duplicate, and it was found that repeat experiments agreed almost exactly.

Part I.—Results of Tests on Carbon and Nickel Steels.

The steels tested in this section consisted of plain carbon steels, with a carbon content ranging from 0·21 to 0·40 per cent. and containing less than 0·16 per cent. of nickel ; and a series of nickel steels, those marked *A* and *AI* having 0·26 per cent. of carbon and 0·43 per cent. of nickel, whilst steels *C*, *D*, and *DI* had approximately 0·55 per cent. of carbon and 1·10 per cent. of nickel. All the steels had about the same manganese content and grain-size, but, owing to variations in the heat treatments applied, the micro-constituents and mechanical properties varied among the respective steels. Photomicrographs of some of the steels are given in Figs. 16 to 21 (Plate III.), whilst the analyses and ultimate tensile strengths are recorded in Tables I. and II. The results of the penetration tests at 250° C. and 350° C. are tabulated in Tables III. and IV., and the effects of variation in the amount of free ferrite are indicated graphically in Figs. 5 to 7. It should be emphasised, however, that in the graphs portraying the influence of free ferrite, whether present as a grain boundary network or intermingled with sorbite or pearlite, no attempt has been made to give an accurate quantitative representation. Each graph should be regarded as a manifestation of the effect of ferrite rather than as a quantitative survey.

In discussing the results shown in the above-mentioned tables and throughout the paper, reference is made to soldered samples only, the unsoldered samples being used chiefly in the determination of the ultimate stress ratio (*see* footnote, Table III.).

TABLE I.—*Carbon Steels.*

Symbol.	Analysis.				Maximum Tensile Stress. Tons per sq. in.
	C. %	Mn. %	Ni. %	Cr. %	
B	0·395	..	0·01	0·14	47·0
CA	0·25	0·59	0·10	0·03	33·8
E	0·21	0·50	0·09	0·03	28·8
F	0·23	0·56	0·15	0·06	30·6
FQ	0·23	0·56	0·15	0·06	47·5

TABLE II.—*Nickel Steels.*

Symbol.	Analysis.				Maximum Tensile Stress. Tons per sq. in.
	C. %	Mn. %	Ni. %	Cr. %	
A	0·26	0·65	0·43	0·06	38·0
AI	0·26	0·65	0·43	0·06	39·2
C	0·53	0·63	1·00	0·06	45·8
D	0·57	0·55	1·20	..	62·4
DI	0·57	0·55	1·20	..	53·4

TABLE III.—*Carbon Steels.*

Symbol.	Testing Temperature, 250° C.				
	Curve Area.	Ultimate Stress Ratio. %	Deflection to Fracture. In.	Yield Load. Kg.	Max. Load. Kg.
B	6·45	88·37	0·1850	360	950
CA	11·41	93·75	0·3950	260	750
E	9·43	98·6	0·3650	175	700
F	9·84	98·6	0·4000	175	700
FQ	11·12	90·0	0·3175	350	900
Symbol.	Testing Temperature, 350° C.				
	Curve Area.	Ultimate Stress Ratio. %	Deflection to Fracture. In.	Yield Load. Kg.	Max. Load. Kg.
B	4·55	80·0	0·1675	300	700
CA	3·69	80·0	0·2875	250	500
E	2·40	77·59	0·1550	170	450
F	2·18	75·0	0·1425	170	450
FQ	4·10	76·47	0·1700	250	650

$$\text{Ultimate Stress Ratio} = \frac{\text{Max. Load with White Metal present}}{\text{Max. Load with White Metal absent}} \times 100.$$

(*A*) *Effect of Free Ferrite.*

The results at 250° C. on the quenched and tempered carbon steels *B*, *FQ*, and *CA*, arranged in order of increasing ferrite,

TABLE IV.—*Nickel Steels.*

Symbol.	Testing Temperature, 250° C.				
	Curve Area.	Ultimate Stress Ratio. %	Deflection to Fracture. In.	Yield Load. Kg.	Max. Load. Kg.
A	7·30	89·5	0·2320	285	850
AI	6·65	79·0	0·2350	285	750
C	1·97	66·7	0·0755	290	700
D	3·76	83·3	0·1100	400	1000
DI	2·37	70·5	0·0875	350	775
Symbol.	**Testing Temperature, 350° C.**				
	Curve Area.	**Ultimate Stress Ratio. %**	**Deflection to Fracture. In.**	**Yield Load. Kg.**	**Max. Load. Kg.**
A	1·85	68·8	0·085	250	550
AI	2·36	71·0	0·1125	250	550
C	1·30	64·7	0·0650	200	550
D	2·13	73·7	0·0800	300	700
DI	2·60	75·6	0·1125	250	605

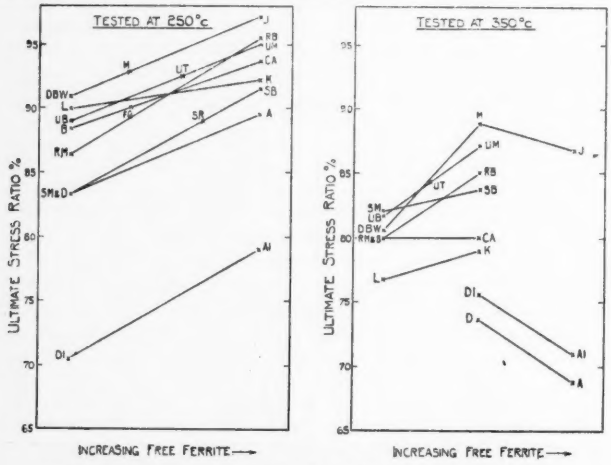

FIG. 5.—Effect of Free Ferrite on Ultimate Stress Ratio.

show that increasing amounts of free ferrite were accompanied by an increase in the ultimate stress ratio, maximum deflection and area enclosed by the load-deflection curve. The tests conducted at 350° C., however, show that there was no appreciable benefit to be derived from increasing amounts of free ferrite.

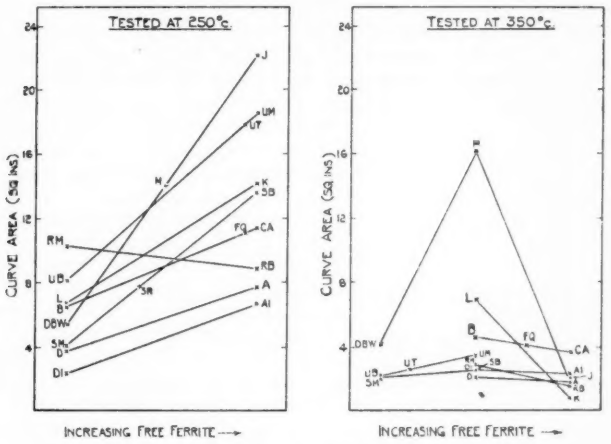

FIG. 6.—Effect of Free Ferrite on Load-Deflection Curve Area.

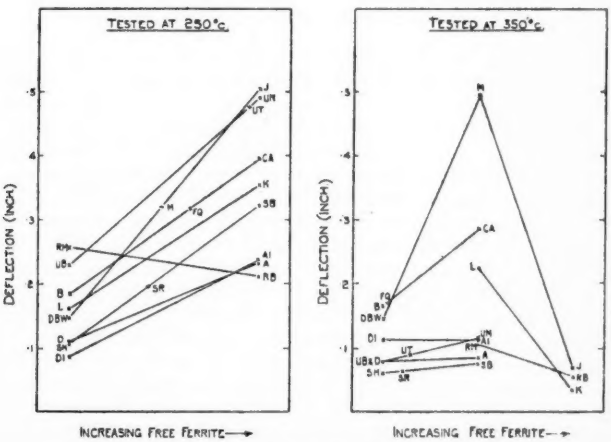

FIG. 7.—Effect of Free Ferrite on Deflection to Fracture.

Although an increase of deflection was obtained on steel *CA* the curve area was least for that material. Comparing the oil-hardened and tempered nickel-steel samples *A* with *D*, and *AI* with *DI*, it is apparent that when tested at 250° C. test-pieces *A* and *AI*, having more free ferrite present, showed an appreciable increase in the ultimate stress ratio, maximum deflection and area enclosed by the respective curves. Also, comparing *A* with *AI* and *D* with *DI* at 250° C., samples having the free ferrite in the form of a network (Plate III.) showed a pronounced increase in those properties. Figs. 5, 6 and 7 indicate that when tested at 350° C., the resistance of the four steels to interpenetration was in the reverse order of their free ferrite contents. Contrary to the conclusion drawn in the tests at 250° C., at 350° *C* there seemed to be no advantage in having the free ferrite in the form of a network (Plate III.). Comparing *C* (normalised) with *DI*, and *F* (normalised) with *FQ* and *CA*, the free ferrite in the hardened and tempered samples *DI*, *FQ*, and *CA*, although less in quantity than in the corresponding normalised specimens, was not in such large angular masses, however, being in more or less continuous membranes. The advantage at 250° C. of this latter formation of the ferrite is shown by the values for the respective curve areas recorded in Tables III. and IV. The evidence appears to show, therefore, that both in the carbon steels and the nickel steels at 250° C. and 350° C., a normalised structure is less resistant to interpenetration than a hardened and tempered structure.

(*B*) *Effect of Variations in Mechanical Properties.*

Comparison of the figures for the yield point, percentage elongation and percentage reduction of area indicates conclusively that there was no definite relationship between those properties and the resistance of the steels to penetration, either at 250° C. or at 350° C. The tensile maximum stress, considered alone, did not bear a definite relationship to any of the interpenetration-resisting properties. The results obtained on samples *A* and *AI*, of practically the same maximum stress, and also on *B* and *FQ*, tested at 250° C., are examples of possible variations in this respect, whilst steel *D*, having a much greater tensile maximum stress than *DI*, gave results at 350° C. intimating that *DI* was more resistant to penetration than *D*. Similarly, steel *B* had a slightly lower maximum stress than *FQ* (Table III.), but the penetration tests at 350° C. indicated that the ultimate stress ratio and the curve area were greater for that steel than for *FQ*. It is possible, however, that if any one steel were heat-treated to varying tonnages, keeping the amount and formation of the ferrite as constant as possible, the tensile maximum stress

would give an indication of the resistance of the steel to the interpenetration of molten white metals.

From the foregoing test results, both at 250° C. and at 350° C., it is evident that the presence of nickel, even in small quantities considerably reduced the resistance of the steel to interpenetration in comparison with plain carbon steel.

PART II.—RESULTS OF TESTS ON NICKEL-CHROMIUM STEELS.

The samples tested in this section of the investigation were machined from oil-hardened and tempered steels, in which the nickel content varied between the limits of 3·41 and 3·71 per cent. There was, however, a wide variation in the carbon and chromium contents, grain-size and quantity of free ferrite. The analyses and ultimate tensile strengths of the steels are recorded in Table V., and the results of the penetration tests are given in Table VI., whilst photomicrographs indicative of the structures of some of the samples are reproduced in Figs. 22 to 27 (Plate IV.).

TABLE V.—*Nickel-Chromium Steels.*

Symbol.	Analysis.				Maximum Tensile Stress. Tons per sq. in.
	C. %	Mn. %	Ni. %	Cr. %	
DBW	0·375	0·46	3·41	0·22	52·6
G	0·20	0·29	3·58	0·53	48·6
H	0·20	0·29	3·58	0·53	48·6
J	0·12	..	3·64	1·11	50·3
K	0·26	0·41	3·43	0·60	50·4
L	0·29	..	3·54	0·54	56·45
M	0·285	..	3·71	0·38	52·55

(*A*) *Effect of Intermingled Ferrite.*

Comparison of the figures given in Table VI. for large grain-size samples *L* and *K*, and small grain-size samples *DBW*, *M*, and *J*, each series being arranged in order of increasing amounts of free ferrite, indicates that with more intermingled ferrite present, the greater was the ultimate stress ratio, maximum permissible deflection and curve area, when tested at 250° C. The increase in the deflection and curve-area was particularly pronounced in the small-grained samples, but, contrary, to the conclusions drawn when considering the carbon and nickel steels, the results obtained on steel *DBW* infer that the presence of ferrite in the form of membranes was not an advantage in oil-hardened and tempered nickel-chromium steels. The test figures at 350° C., given in Table VI., and depicted in Figs. 5 to 7,

show that for the large grain-size samples *L* and *K*, both the deflection and curve area were higher in the case of the steel *L*. The low deflections and curve areas obtained on steels *K* and *G* indicate the possibility that when stressed at 350° C. in the presence of molten metal, there is an optimum amount of intermingled ferrite which will have a beneficial effect, and that the presence of more intermingled ferrite than this amount is deleterious.

Dealing next with the test results at 350° C. on the small-grained samples *DBW*, *M* and *J*, it is apparent that the sample

TABLE VI.—*Nickel-Chromium Steels.*

Symbol.	Testing Temperature, 250° C.				
	Curve Area.	Ultimate Stress Ratio. %	Deflection to Fracture. In.	Yield Load. Kg.	Max. Load. Kg.
DBW	5·42	90·9	0·1475	500	1000
G	5·90	90·7	0·1575	400	975
H	4·02	87·0	0·1125	400	935
J	22·18	97·2	0·5050	500	1020
K	14·09	92·2	0·3550	600	950
L	6·65	90·0	0·1615	600	1080
M	13·95	92·8	0·3200	600	1090
Symbol.	Testing Temperature, 350° C.				
	Curve Area.	Ultimate Stress Ratio. %	Deflection to Fracture. In.	Yield Load. Kg.	Max. Load. Kg.
DBW	4·17	80·6	0·1475	365	725
G	3·22	78·4	0·1125	350	725
H	3·15	75·7	0·1125	350	700
J	2·06	86·8	0·0700	450	825
K	0·79	79·0	0·0350	475	750
L	6·91	76·8	0·2250	475	730
M	16·18	88·9	0·4950	475	800

DBW again showed the lowest ultimate stress ratio, and also a comparatively low maximum deflection. Sample *M*, having an intermediate amount of intermingled ferrite, gave the best results for the ultimate stress ratio, deflection and curve area, the latter, in fact, being greater than the area enclosed by the load-deflection curve obtained when testing that particular steel at 250° C. Increasing the amount of intermingled ferrite, however, resulted in a very marked fall in the ultimate stress ratio, deflection and curve area, as shown by the values obtained on sample *J*. It is evident, therefore, that at 350° C., although a small amount of intermingled ferrite was beneficial when

coupled with small grain-size, no appreciable benefit was derived from its presence when in conjunction with a large-grained microstructure; also, amounts of intermingled ferrite above a certain optimum quantity distinctly reduced the resistance of the steel to interpenetration, irrespective of the grain-size.

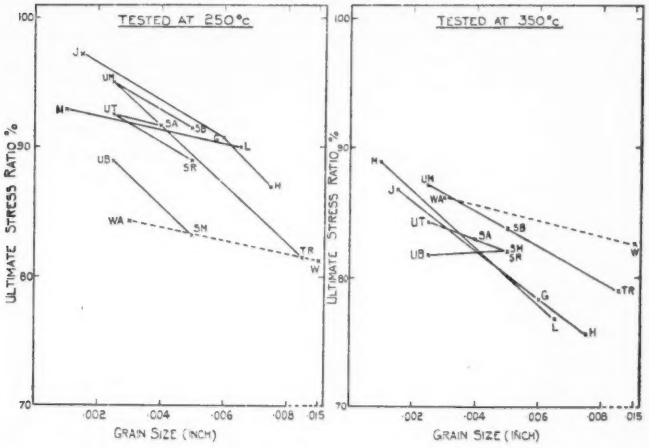

FIG. 8.—Effect of Grain-Size on Ultimate Stress Ratio.

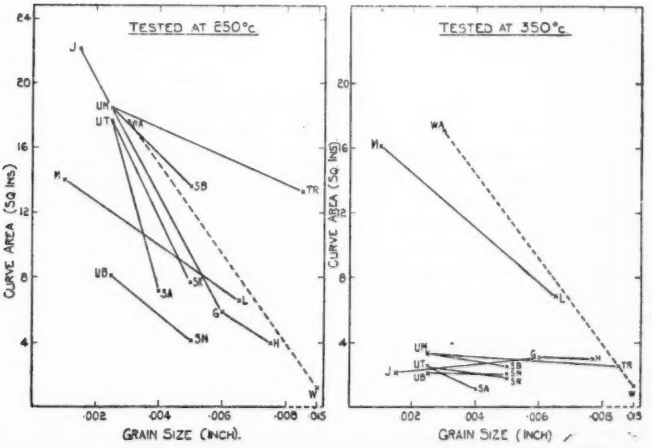

FIG. 9.—Effect of Grain-Size on Load-Deflection Curve Area.

(B) Effect of Grain-Size.

Comparison of the test results obtained for steels *G* and *H*, and also for steels *M* and *L*, the first in each pair having an appreciably smaller grain-size (Plate IV.), shows that, in steels otherwise very similar, a large grain-size resulted in an appreciable reduction in the resistance to interpenetration, Figs. 8 to 10 giving ample corroboration of that conclusion. The beneficial results obtained by a combination of small grain-size and fine sorbite with a small amount of intermingled ferrite, are illustrated by reference to the test values obtained on steel *M*. The effects

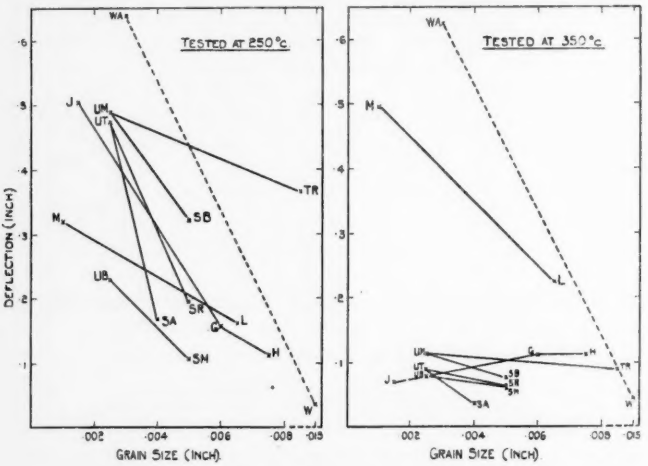

FIG. 10.—Effect of Grain-Size on Deflection to Fracture.

of grain-size were similar at 350° C. to those indicated for the respective steels when tested at 250° C., but were less marked at the higher temperature. The position of steel *J* in Figs. 9 and 10 (350° C. tests), with reference to the larger-grained steels *G* and *H*, constitutes the only exception, and the relatively low values obtained on the former steel at this temperature are attributed to the possibility that the deleterious influence of the large amount of free ferrite had more than nullified the beneficial effects of the small grain-size.

PART III.—RESULTS OF TESTS ON NICKEL-CHROMIUM-MOLYBDENUM STEELS AND AN AUSTENITIC STEEL.

The analyses and ultimate tensile strengths of the respective steels are given in Table VII., whilst in Table VIII. the results

of the penetration tests are recorded. All samples were oil-quenched and tempered, except steels *RM* and *RB* which were annealed, *V* and *VMo* which were air-hardened and tempered, *W* which was tested in the as-forged condition, and *WA* which was water-quenched.

TABLE VII.—*Nickel-Chromium-Molybdenum and Austenitic Steels.*

Symbol.	Analysis.						Maximum Stress. Tons per sq. in.
	C. %	Mn. %	Ni. %	Cr. %	Mo. %	W. %	
N	0·33	0·46	1·55	0·59	0·25	..	53·3
O	0·315	0·44	3·38	0·69	0·24	..	64·6
DGW	0·31	0·50	3·49	0·11	0·30	..	58·4
RM	0·28	0·39	2·49	0·58	0·37	..	50·8
RB	0·28	0·39	2·49	0·58	0·37	..	50·0
SA	0·26	0·44	2·55	0·58	0·42	..	56·4
SM	0·28	0·39	2·54	0·51	0·40	..	54·5
SR	0·28	0·39	2·54	0·51	0·40	..	52·2
SB	0·28	0·39	2·54	0·51	0·40	..	51·4
TR	0·23	0·42	2·51	0·72	0·43	..	50·8
UB	0·275	0·48	2·89	0·57	0·22	..	46·2
UT	0·275	0·48	2·89	0·57	0·22	..	44·4
UM	0·275	0·48	2·89	0·57	0·22	..	44·4
VX	0·285	0·71	2·45	0·60	0·64	..	62·9
VY	0·24	0·55	3·31	0·63	0·58	..	65·1
V	0·27	0·41	3·72	1·10	0·05	..	58·6
VMo	0·31	0·64	3·96	1·36	0·24	..	65·8
W	0·63	7·34	9·26	4·12	0·63	0·48	57·4
WA	0·63	7·34	9·26	4·12	0·63	0·48	50·7

(*A*) *Effect of Intermingled Ferrite* (*Figs.* 5, 6 *and* 7).

Comparing samples *SM*, *SR* and *SB* of very large grain-size, and *UB*, *UT* and *UM* of small grain-size, each series being arranged in order of increasing amounts of intermingled ferrite (*see* Figs. 28 to 33, Plate V.), the evidence indicates that at 250° C., increasing amounts of intermingled ferrite resulted in a pronounced increase in the ultimate stress ratio, deflection and curve area. The effect of the intermingled ferrite was more pronounced and also more beneficial with small grain-sizes. The results obtained on the annealed steels *RM* and *RB*, *RB* having a preponderance of free ferrite, did not conform with this conclusion, the apparent discrepancy being due in all probability to the fact that the free ferrite was in large areas and not intermingled in a finely dispersed form. Similar comparisons of the test values obtained at 350° C. lead to the conclusion that no appreciable effect was caused by an increase of intermingled

ferrite in the large-grained samples, but when the specimens were of small grain-size, increasing the amount of *intermingled* ferrite resulted in a slight increase in the penetration-resisting properties of the steels. The results on specimens *RM* and *RB* confirmed previous observations, *viz.*, that too much free ferrite reduced the penetration-resistance of the stressed steels, par-

TABLE VIII.—*Nickel-Chromium-Molybdenum and Austenitic Steels.*

Symbol.	Testing Temperature, 250° C.				
	Curve Area.	Ultimate Stress Ratio. %	Deflection to Fracture. In.	Yield Load. Kg.	Max. Load. Kg.
N	9·32	95·1	0·2200	525	1070
O	5·93	94·1	0·1250	625	1200
DGW	4·73	90·2	0·1125	650	1150
RM	9·30	86·4	0·2575	450	950
RB	8·85	95·5	0·2125	450	1050
SA	7·20	91·7	0·1700	600	1100
SM	4·10	83·3	0·1075	550	1000
SR	7·69	89·0	0·1950	500	1015
SB	13·61	91·5	0·3225	500	1025
TR	13·32	81·5	0·3650	480	880
UB	8·08	89·0	0·2300	450	890
UT	17·71	92·5	0·4750	450	925
UM	18·50	95·0	0·4900	450	950
VX	9·11	90·2	0·2000	675	1150
VY	11·70	91·0	0·2525	700	1160
V	3·81	93·2	0·0925	600	1100
VMo	5·78	93·3	0·1275	650	1190
W	1·21	81·3	0·0375	650	1000
WA	17·68	84·3	0·6400	300	750

Symbol.	Testing Temperature, 350° C.				
	Curve Area.	Ultimate Stress Ratio. %	Deflection to Fracture. In.	Yield Load. Kg.	Max. Load. Kg.
N	2·25	87·4	0·0625	450	900
O	2·99	89·4	0·0800	550	1050
DGW	3·05	88·2	0·0900	430	900
RM	2·92	80·0	0·1075	400	750
RB	1·57	85·0	0·0575	400	850
SA	1·23	83·0	0·0375	475	900
SM	2·10	82·1	0·0625	480	850
SR	1·94	82·0	0·0630	460	820
SB	2·61	83·8	0·0775	460	830
TR	2·60	79·0	0·0900	450	750
UB	2·15	81·7	0·0800	385	735
UT	2·60	84·3	0·0900	400	750
UM	3·45	87·1	0·1150	400	775
VX	3·45	86·4	0·0950	550	950
VY	3·43	87·0	0·0925	600	1000
V	2·08	84·1	0·0625	450	925
VMo	1·61	86·2	0·0450	600	1000
W	1·37	82·6	0·0450	550	950
WA	17·13	86·2	0·6250	175	750

ticularly when the free ferrite was not in a finely and uniformly dispersed form.

(B) Effect of Grain-Size.

The series of steels included in this section of the investigation offered a number of comparisons on the effect of grain-size, and these have been shown in graph form in Figs. 8 to 10. Photomicrographs illustrating some of the differences in grain-size are reproduced in Figs. 28 to 33. As a result of the comparisons it

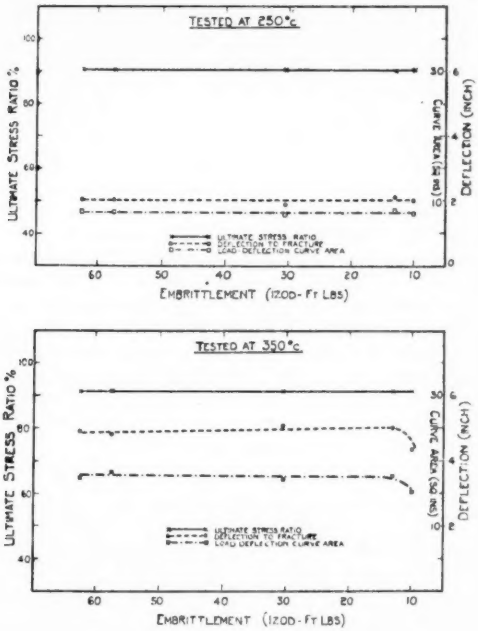

FIG. 11.—Effect of Embrittlement.

can be concluded, quite definitely, that at 250° C. large grain-size resulted in an appreciable decrease in the resistance of the materials to the penetration of molten white metals. Similar indications were observed when testing at 350° C., but at this temperature the beneficial effect of small grain-size was much less pronounced, except in the case of the austenitic steel specimens *W* and *WA* (Figs. 13 and 14, Plate I.). The results of the tests on the latter two samples, however, showed that in this type of

steel large grain-size was a distinct disadvantage at both temperatures.

Part IV.—The Effect of Temper-Brittleness.

For these tests, samples from material *K*, the analysis of which is given in Table V., were oil-hardened and tempered to have a maximum tensile strength of 50 tons per sq. in. Samples of the material were then reheated to 550° C., and cooled from that temperature at predetermined rates to give progressively lower toughness, as measured by the Izod impact test. The impact figures varied from a mean Izod value of 62·5 ft.-lb., down to 10 ft.-lb. The samples were then tested at 250° C. and 350° C., both with and without solder or bearing metal present, the results being depicted in graph form in Fig. 11. The curves show conclusively that there was no connection between the Izod value and the resistance to penetration, as measured by the values for the ultimate stress ratio, deflection and curve area. Whether tested at 250° C. or at 350° C., the penetration test results were practically the same throughout the range of impact values observed. The microstructure of the steel, in the re-treated condition, was similar to that of the nickel-chromium steel *M*, a photomicrograph of which is reproduced in Fig. 27, and consisted of small grains of sorbite with a small amount of intermingled ferrite, uniformly and finely dispersed. The tests consequently confirm the conclusion drawn from the results on steel *M*, namely, that such a microstructure is particularly resistant to penetration, the resistance being greater at 350° C. than at 250° C.

Summary.

An investigation was carried out to ascertain the effect of variations in temperature, rate of application of stress, mechanical properties, microstructure and chemical composition, on the resistance of stressed steels to the penetration of molten white metals. It was found that premature failure only occurred when the material was stressed in tension in the presence of molten white metal, but the composition of the molten metal was apparently of no significance. The yield and maximum loads withstood by the different steels, in the presence of molten white metal, decreased with increasing temperature, and also with slower rates of loading at constant temperature. The variations of temperature and loading rate, however, were much more profound in their influence on the maximum deflection which the stressed steels would withstand in the presence of molten white metals, there being a pronounced decrease in deflection with slower rates of loading and also with increasing temperatures. The maximum permissible deflection is considered to be of importance, since with

small deflections, even when the maximum load was high, the area of the load-deflection curve, and therefore the energy required to cause fracture, was comparatively small. Series of tests were carried out at 250° C. and 350° C., with and without molten white metal in contact with the surfaces of the steel samples, which were stressed in tension, a constant rate of loading being used; the deflection at the beginning and end of each load increment was measured and then complete load/extension diagrams were plotted for each test. From the results of the tests the following conclusions were drawn:

(1) When testing samples at 250° C. or 350° C. in the presence of molten white metal, the tensile properties of the materials were no reliable indication of the resistance of the steels to interpenetration. For example, although the tensile properties of the series of nickel-chromium steels were fairly constant, very considerable differences were recorded between the different steels in relation to their resistance to the interpenetration of molten metals.

(2) At both temperatures the addition of nickel up to 1·2 per cent. to carbon steels was distinctly deleterious, whilst the results obtained on nickel-chromium steels were either better or worse than those obtained on carbon steels, depending upon the microstructure. The results showed that carbon steels, as a class, were more uniform than the alloy steels in their resistance to failure through interpenetration, but it was possible to heat-treat the nickel-chromium and the nickel-chromium-molybdenum steels so that they had a microstructure which would resist the penetration as well as, and in some cases distinctly better than, the carbon steels. Such a structure consisted of very small grains of fine sorbite with a small amount of intermingled ferrite. The addition of molybdenum to nickel-chromium steels was not beneficial from the point of view of resistance to interpenetration, those steels with molybdenum showing no superiority over similar steels without that element.

(3) When testing nickel steels and carbon steels at 250° C., it was found that increased amounts of ferrite increased the resistance to penetration, particularly when the free ferrite was in the form of membranes round the sorbitic grains, as distinct from angular masses. Quenched and tempered structures therefore gave better results than normalised or annealed structures. When testing at 350° C., however, it was concluded that the beneficial effect of the free ferrite had almost disappeared. When testing the nickel-chromium and the nickel-chromium-molybdenum steels at 250° C., the more ferrite intermingled with the sorbite, the greater was the resistance to penetration, but in this class of steel the existence of a free ferrite network was of no advantage.

As had been found for the nickel steels and carbon steels, when subjected to stress at 350° C., no advantage was gained by increasing amounts of intermingled ferrite above a very small quantity, and, in fact, anything in excess of such a small amount aided the penetration.

(4) The resistance to interpenetration of molten metal, at both temperatures, varied inversely as the grain-size of the material. The effect of grain-size was very pronounced at 250° C., but the benefit accruing from small grain-size was not so marked at 350° C.

(5) A small-grained austenitic steel offered much greater resistance to penetration at both temperatures than any of the other steels, with the exception of a nickel-chromium steel having a small grain-size, very fine sorbite and a small amount of intermingled ferrite.

(6) The resistance of all the steels, except the two referred to in paragraph (5) above, was appreciably less at 350° C. than at 250° C.

(7) The time required at the maximum load for fracture to occur, was considerably less at 350° C. than at 250° C., failure being almost instantaneous in most cases at the higher temperature.

(8) Temper-brittleness, as determined by the Izod impact test, appeared to have no influence on the resistance of the steel to interpenetration.

In conclusion, the author wishes to thank the Directors of the English Steel Corporation, Vickers Works, Sheffield, for granting permission to publish the contents of this paper; Mr. J. H. Slagg, for the preparation of the photomicrographs; and Mr. P. H. Frith, for his assistance during the testing of the materials.

CORRESPONDENCE.

Dr. L. J. G. VAN EWIJK (Materials Department, National Institute for Aeronautical Research, Amsterdam) wrote that Mr. Goodrich's paper had interested him very much, since it dealt with the same subject as his (Dr. van Ewijk's) paper read before the Institute of Metals at the Annual Meeting of this year.

Mr. Goodrich studied the phenomenon of penetration on many types of steel (carbon, nickel, nickel-chromium, nickel-chromium-molybdenum steels) and came to a series of conclusions which were very noteworthy. Like other investigators, he found that premature failure only occurred when the material was stressed in tension in the presence of molten white metal, and that the composition of the molten metal was apparently of no significance.

He stated that the tensile properties of the steels formed no reliable indication of the resistance to penetration, and also that the temper-brittleness, as determined by the Izod impact test, appeared to have no influence. In the tests carried out by the writer a certain relationship between the Izod value and the resistance to penetration could be detected, but he thought that the connecting link between these two properties was formed by the grain-size; coarse-grained nickel-chromium steel showed a low resistance to penetration as well as a low impact value.

Mr. Goodrich found that it was possible to heat-treat the nickel-chromium and the nickel-chromium-molybdenum steels so that they possessed a resistance to penetration equal to or even better than that of carbon steels. In Tables VI. and VII. the maximum stress of the steels investigated was mentioned; it ranged from 48·6 to 56·5 for the nickel-chromium, and from 44·4 to 65·8 tons per sq. in. for the nickel-chromium-molybdenum steels.

Dr. van Ewijk asked whether Mr. Goodrich had investigated Cr–Ni or Cr–Ni–Mo steels of higher tensile strength, say, 100 tons per sq. in. (*e.g.*, B.S. Specification 2.T.2) and whether he thought that it would be possible to decrease the liability to penetration sufficiently without too much loss of strength.

The rôle ascribed to the free ferrite was not quite clear. Did Mr. Goodrich think that danger of penetration might occur when the ferrite formed angular masses,[1] and was the membrane form therefore more preferable; or were the sorbitic grains to be feared, and did the free ferrite reduce this danger by enveloping the sorbite?

[1] In the discussion on the writer's paper, Dr. Desch said that perhaps the form of the crystals was of more importance than the size, those of sharp polygonal form being more susceptible.

In his opinion all those interested in the problem of the penetration of steel by low-melting-point metals and alloys were very much indebted to Mr. Goodrich for the publication of his results. Although there were still many points to be considered, the paper distinctly meant a step forward in the direction of the solution of the problem.

AUTHOR'S REPLY.

Mr. GOODRICH, in reply, wished to thank Dr. van Ewijk for his interest in, and comments on, the paper, and was pleased to hear that his conclusion confirmed, to a large extent, Dr. van Ewijk's evidence.

The remarks concerning the correlation of the Izod impact value and the resistance to penetration were extremely interesting, and the author agreed that some relationship might ultimately be determined for materials having a very large grain-size, even though the evidence at present available for steels of small or medium grain-size gave no indication of such a relationship.

It might be of interest to Dr. van Ewijk to hear that the coarse-grained steels *SB*, *SM*, *SR*, and *TR*, although offering relatively low resistance to interpenetration, had Izod impact values of over 55 ft.-lb.

With reference to the question of the liability to penetration in the case of nickel-chromium or nickel-chromium-molybdenum steels of approximately 100 tons per sq. in. tensile strength, the author had only a small amount of evidence, upon which it would be improper to base any definite conclusions. That evidence, however, indicated that, even in the presence of small grain-size, such steels were liable to interpenetration unless the composition was adjusted to the minimum carbon, nickel or chromium contents consistent with the production of the requisite mechanical properties.

Such an adjustment of composition, sufficient to ensure the presence of a small quantity of intermingled ferrite even in thin sections of material, appeared to increase the resistance to interpenetration appreciably.

With regard to the question of the formation of the ferrite, it would appear that when the ferrite was present in the form of large angular masses its beneficial influence was much less than would be the case were that same quantity of ferrite present in the form of membranes or uniformly and finely dispersed through the matrix.

It was the author's opinion that sorbitic grains were not to be feared unless they were of large size and completely free from intermingled ferrite.

SUBCUTANEOUS EFFECTS DURING THE SCALING OF STEEL.*

By R. GRIFFITHS, M.Sc. (University College, Swansea).

Summary.

An explanation is advanced for the formation of globules and spots below the steel surface when oxidising under certain conditions.

Introduction.—Modern requirements of surface finish on steel products have necessitated the closest attention being paid to causes of minor defects. This applies particularly in the tinplate industry, where the need for economy has resulted in the thickness of the tin coating being cut down to the minimum commensurate with adequate protection. In this and other fields, minor subcutaneous changes in the structure and composition of the articles during heat treatment may be reflected in the properties of the finished products—in particular, a decreased resistance to fatigue and corrosion may result.

The present paper deals with subcutaneous changes which have been observed during the oxidation of various steels in the laboratory, but which may prove of industrial interest.

Previous Work.—The penetration of oxygen in steel on scaling has been referred to by Stead,[1] who described the structure found at the surface of a steel casing from an ironworks stove. Pfeil,[2] in his valuable paper on the oxidation of iron and steel at high temperatures, mentioned the absorption of oxygen by steel, particularly alloy steels. Closely related to the present author's experiments is one carried out by McCance[3]. A sample of ordinary mild steel was packed in a crucible with crushed mill scale and the whole was heated for 8–12 hr., at a temperature approaching 1,200° C. The explanation suggested for the migration of the oxide into the steel was that FeO is soluble in solid iron at a temperature of 1,200° C. and is less soluble at lower temperatures, so that it is deposited as the iron cools down. Recently, Whiteley[4] discussed the network structure formed at the boundaries of either α or γ surface grains when steel is scaled.

Practical Investigation.—Many iron and steel samples, widely differing in composition, were treated and examined during the

* Received June 16, 1935.

investigation, but reference to those cited in Table I. only will be made in particular.

TABLE I.—*Chemical Analysis of Samples.*

Material.	C. %	Si. %	S. %	P. %	Mn. %	Cr. %	Al. %	Ni. %
Pure Iron . .	..	..	..	..	..	..	..	0·02
Remelted electrolytic iron . .	Trace	Nil	0·005	0·003	Trace	..	..	..
Armco iron . .	0·021	Trace	0·025	0·005	0·041	..	..	..
Carbon steel	0·53	0·044	0·036	0·040	0·50	..	..	..
Chromium steel .	0·43	0·035	0·029	0·024	0·05	0·91	..	..
Aluminium steel .	0·019	0·02	0·028	0·008	0·038	..	1·01	..
Manganese steel .	0·07	0·098	0·046	0·022	0·95	..	..	..
Tin bar . .	0·06	0·035	0·03	0·02	0·28	..	Trace	..

Briefly, the experiments consisted of subjecting the steels to slow oxidation at certain temperatures for periods ranging from one hour to three weeks.

Holes, $\frac{3}{16}$ in. in diam., were drilled in the specimens, which were in the form of cylinders 1 in. in diam. by 1 in. long. Finely-ground scale, obtained by the oxidation of electrolytic iron, was rammed into the cavity and then sealed by means of a tightly-fitting screw plug.

Using an electrically-heated silica-tube furnace, different specimens were then subjected to three conditions of heating at the various temperatures for the various times. Under the first condition, the specimens were heated *in vacuo*. In the second series, slow oxidation of the exterior of the specimens was allowed in a limited supply of air, and in the final series, to produce a carburising effect, the specimens were heated in coal gas.

Each specimen was then sectioned transversely through the rod to give a section with the oxide-filled cavity at the centre. The specimens were then mounted to preserve the exterior surfaces, for which purpose sealing-wax, low melting-point alloy or bakelite was used. In some cases the exterior scale was carefully removed before mounting.

Examination of the Specimens.—In the series heated at a temperature of 950° C. *in vacuo* for three weeks, it was found that in the Armco iron, carbon, aluminium, manganese and chromium steels, greyish-black globules or areas were present near the metal-scale interface, together with finer spots which extended into the steel.

Plate VI. illustrates some typical appearances of three of the steels at the affected areas. Fig. 1 shows that of the manganese steel, Fig. 2 the aluminium steel and Fig. 3 the chromium steel, all in the unetched condition. It will be seen that the appearance of the particles differs somewhat in the various steels.

SUBCUTANEOUS SURFACES OF OXIDISED STEELS.

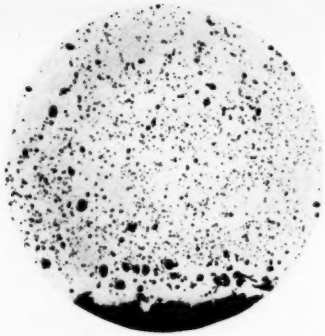

FIG. 1.—Manganese Steel. Surface near scale-filled cavity.

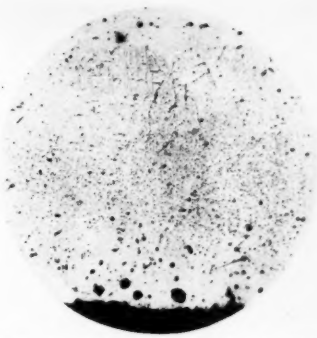

FIG. 2.—Aluminium Steel. Surface near scale-filled cavity.

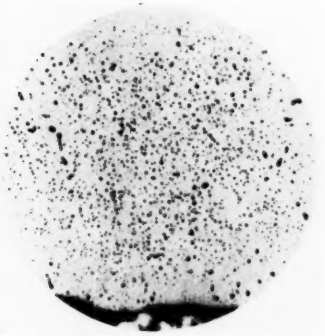

FIG. 3.—Chromium Steel. Surface near scale-filled cavity.

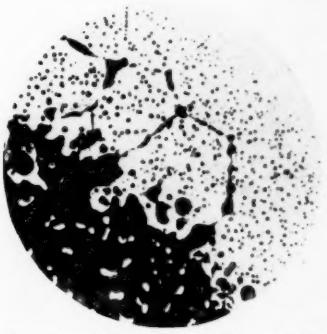

FIG. 4.—Chromium Steel. Surface oxidised in air.

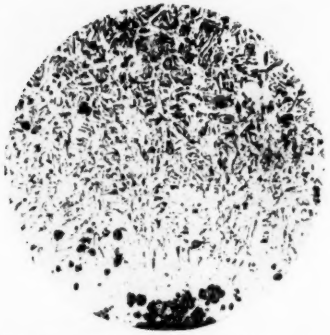

FIG. 5.—Carbon Steel. Surface near scale-filled cavity.

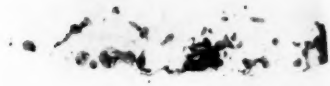

FIG. 6.—Tin Bar. Surface slowly oxidised in air.

(Magnification, 50 diameters, reduced in reproduction to half-size linear.)

Griffiths.

To face p. 68.

All the steels which contained carbon previous to the experiments were found to be decarburised completely as a result of the heating for three weeks *in vacuo*. No change was observable in the appearance of the outer surface of the specimens, from the point of view of the formation of globules and spots.

Turning now to the series exposed to a slowly oxidising atmosphere for seven days at 925° C., it was found that the size of the individual globules was larger, but the thickness of the affected area of the steel was somewhat less at the outer scale-steel interface than at the scale-filled cavity surface. This is illustrated clearly in Fig. 4, which shows a chromium steel of similar composition to that represented in Fig. 3.

As in the series heated *in vacuo*, the appearance of the globules and spots in the various steels differed considerably. During the seven days' heating, decarburisation had taken place to a limited extent in those steels containing carbon. This is illustrated in Fig. 5, which shows the area near the inside scale-filled cavity of the 0·53 per cent. carbon steel. It will be noted that globules and spots appear in the decarburised area.

With the carbon steel, when it was heated for ten days at 950° C., it was found that an inner ring of ferrite free from particles was formed in the scale-filled cavity, due probably to the reduction of the iron oxide in the cavity. The steel in the vicinity of the cavity contained globules and spots as in the previous cases.

Fig. 6 shows the effect obtained on heating a piece of tin bar for 12 hr. at 950° C. in air. It will be seen that irregular globules and a few spots have been formed. On heating for shorter periods, such as 4 hr., the affected area was not so pronounced but was still appreciable. The effect was observable at temperatures as low as about 850° C., although naturally the degree of penetration was correspondingly smaller than at higher temperatures.

Remelted electrolytic iron revealed affected areas to a small extent, and even an iron containing 0·02 per cent. of nickel only as impurity showed distinct globules, after immersion in scale and heating for 12 hr. at 950° C.

Finally, in the case of the steels heated in a carburising atmosphere for fifteen days, the globules and spots were present in the vicinity of the oxide-filled cavity as in the previous cases, but to a somewhat smaller extent than in the air-oxidised series. Carburisation of the steels had, of course, taken place, but in addition some of the scale in the cavity had been reduced to iron and become carburised. It was difficult to find the line of demarcation between the original wall of the scale-filled cavity and the steel.

Consideration of Results.—The experiments show that by heating steels, which contain certain elements, in contact with scale, globules and spots are formed within the steels. A similar

result is obtained by oxidising at a certain rate to form scale on the surface of the steels themselves ; but the globules and spots are not obtained when the steels alone are heated *in vacuo*.

It thus appears that oxygen is necessary for the formation of the globules and spots. A sufficient supply of oxygen, such as is obtained by certain rates of progressive surface oxidation, appears to increase the size of the individual particles in comparison with that obtained with a limited supply of oxygen as in a scale-filled cavity. The oxygen supplied by the scale in the cavity would be obtained during the initial stages from such oxides as Fe_2O_3, and Fe_3O_4, and during the subsequent stages from FeO, which would cause the main effect.

The character of the globules and spots differs according to the elements contained in the steel.

The author would advance the following explanations to account for the phenomena described. The oxygen dissociating from the ferrous-oxide scale within the cavity enters the steel, tending to saturate the surface layers with oxygen in solid solution. Some of the oxygen, however, reacts with those elements in the steel which have a higher affinity for oxygen than has iron. Such elements are chromium, manganese, silicon, aluminium, &c.

It can be imagined that an oxygen atom entering the steel, in its passage through the crystals, comes in contact with, to take an example, an atom of manganese, with which it combines and precipitates as manganese oxide. Further atoms of manganese and oxygen may subsequently diffuse towards this and enlarge it. In addition, atoms of iron may possibly be dissolved in the particle, as well as atoms of other elements, such as silicon, forming in such cases a complex globule or area.

This precipitation occurs in the first place at random at a number of nuclei in the steel and possibly at the boundaries of the crystals. When progressive oxidation of the steel surface is taking place simultaneously, there will be a removal of the affected area of steel by its conversion into scale. It is conceivable that the rate of scale formation can equal, if not exceed, that of oxygen penetration and precipitation, but the rates of oxidation in air commonly encountered are not fast enough for the latter action to be completely overcome. It would appear that the best conditions for maximum precipitation would be the least possible surface attack of the steel commensurate with an adequate supply of oxygen to precipitate the oxidisable elements in the steel.

Conclusions and Applications.—Whenever scaling of an article occurs—for instance, during heat treatment—particularly if the surface oxidation is slow, the formation to some extent of globules and spots is to be expected. Whether the affected area is removed by subsequent processes, such as pickling or machining, will depend,

inter alia, on the extent of the affected area and the amount of metal removed.

It would also appear that the more oxidisable the elements contained in the steel the greater is the extent of precipitation and formation of globules and spots, so that where a specially good surface is required the steel should contain a minimum of elements having a high affinity for oxygen—that is, any excess of deoxidising agents, such as aluminium, &c., should be at a minimum.

Unless the affected area is removed completely, the effects of the phenomenon on a steel intended to be fatigue-resisting may be to initiate surface cracks. In tinplate manufacture, whilst the affected area, if any, must be extremely thin, owing to the rolling processes, &c., subsequent to its formation, its possible presence cannot be entirely disregarded. It is conceivable that its effect would be to increase the possibility of porosity, particularly if the sheet is to be used for deep stamping. The surface of the sheet containing the non-metallic particles might not be able to withstand the stressing occurring in the process, and minute incipient cracks might be formed at the surface of the steel. The support of the tin coating would not then be continuous, so increasing the liability to failure and the resulting porosity of the coating.

The author would like to express his thanks to Principal C. A. Edwards, F.R.S., for facilities for carrying out the work.

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(1) J. E. Stead: *Journal of the Iron and Steel Institute*, 1921, No. I. p. 271.
(2) L. B. Pfeil: *Journal of the Iron and Steel Institute*, 1929, No. I. pp. 501–560.
(3) A. McCance: *Transactions of the Faraday Society*, 1925, vol. 21, Part 2, p. 189.
(4) J. H. Whiteley: *Journal of the Iron and Steel Institute*, 1935, No. I. p. 181.

CORRESPONDENCE.

Mr. A. B. WINTERBOTTOM (Trondheim, Norway) wrote that the author's explanation of his interesting observations might be elaborated a little in accordance with existing information concerning Fe–O equilibria and scaling. Thus if a diagram (Fig. A) was drawn representing a section through the three oxide phase layers and the steel base, oxygen concentrations for the four

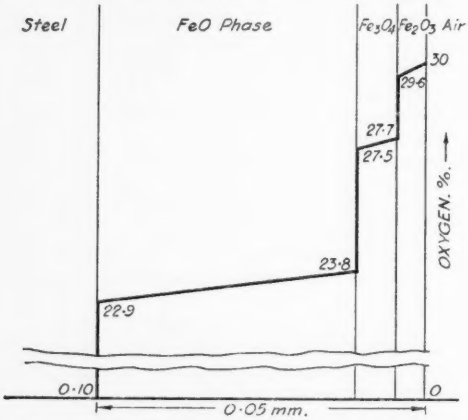


FIG. A.—Diffusion Gradients after 15 min. Scaling at 900° C.

two-phase equilibria could be marked on the interface verticals to an appropriate scale. Clearly, diffusion within each layer destroyed the equilibria, with the result that transfer of iron and oxygen across the interfaces occurred and these moved in the opposite direction to the iron diffusion. So long as a magnetite layer was present in contact with the FeO phase that must tend to saturation with oxygen, *i.e.*, unsaturation with respect to iron, and therefore iron diffused out from the steel, the boundary moved in, and scaling proceeded. If only FeO phase saturated with iron was in contact with the steel, scaling ceased but diffusion of oxygen into the steel continued. An atmosphere in equilibrium with iron and the FeO phase should therefore give the maximum rate of oxygen diffusion inwards and of subcutaneous effect, as was observed.

Turning to the actual mechanism of the subcutaneous precipitation, it seemed reasonable to infer that the oxygen concentration in equilibrium with the precipitated oxides was determined by the element concentrations themselves, *i.e.*, that precipitation

was conditioned by the oxygen concentration exceeding a value given by what might be termed an oxide solubility product $[X]\,[O] < [Fe]\,[O]$. A sufficient concentration of an element with a low solubility product should therefore suppress the effect by limiting it to the surface. It would be interesting to know whether the effect could be produced in the higher alloy steels, such as stainless iron, for example, where good finish was also important.

Mr. K. G. Lewis (Wolverhampton) wrote to congratulate Mr. Griffiths on his paper. To his mind the results appeared somewhat disturbing, and might account for inclusions occurring, principa ly near the surface, in what was apparently good steel in the cast condition.

He would be glad if the author would clarify the meaning of the sentence " No change was observable in the appearance of the outer surface of the specimens, from the point of view of the formation of globules and spots." It was difficult to visualise the presence of those globules, presumably the products of oxidation, in zones where the vacuum created would surely be exerting a much greater effect.

With regard to Mr. Griffiths' comparison of Figs. 3 and 4, it was interesting to note the respective depths of oxidation. While the depth of oxidation was certainly greater in the neighbourhood of the scale-filled cavity, the " degree of oxidation " appeared to be similar—perhaps greater—near the surface oxidised by air. Under the conditions of experiment the greater depth of penetration would be expected, since oxygen formed in the cavity must be under pressure and would thus tend to be forced further into the steel. The reason for the much larger type of globule (Fig. 4) obtained by oxidation in air, was, however, obscure. The writer would ask if the sites of those various globules or inclusions were determined by etching—particularly in the case of Fig. 4. The magnification given ($\times$ 50) was low, and in such circumstances it was difficult to imagine their disposition. Apparently they did not occur at grain boundaries. It would be of interest to know whether they were distributed throughout the crystal grains, or whether they occurred haphazardly both in the grain and at the boundary. The author's explanation of the reaction of the oxygen with elements in the steel appeared to satisfy the facts, since very little in the nature of a grain boundary, where oxide was frequently found in cases of over-oxidation, was detectable from his micrographs. A rate of scaling sufficient to remove a slightly defective surface was a fairly common occurrence in heat treatment.

The danger of incipient cracks was well emphasised by Mr. Griffiths' work. Pickling, sand-blasting, turning and dressing

were in the main expensive operations. U. R. Evans and the writer[1] had pointed out the advantages of scale removal before applying protective coatings. When the exposed surface has to be "cleared" in addition, mechanical methods must generally be adopted and the extra expense could be considerable. Mr. Griffiths' findings offered an interpretation of apparently inexplicable results, *e.g.*, in forgings and some welding processes where heat treating had to be carried out repeatedly—and very often slowly—before the work was completed. In the writer's opinion inclusions which were not rolled-in scale were frequently encountered in the neighbourhood of laps. The method of experiment adopted in the author's work—by heating in contact with oxide in an enclosed space—might offer a possible explanation of that defect, particularly on material which had been subjected to further heat treatment and rolling or forging.

AUTHOR'S REPLY.

The AUTHOR, in reply, wrote that he was interested in the observations of Mr. Winterbottom and Mr. Lewis. Interesting and valuable results should be obtained from work carried out on the suggestion of Mr. Winterbottom.

With regard to Mr. Lewis's query on the sentence "No change was observable in the appearance of the outer surface of the specimens, from the point of view of the formation of globules and spots," that merely meant that no globules and spots of the type under discussion could be discerned at the outer surface of the specimens which had been heated *in vacuo*.

The author rather doubted the possibility of any appreciable pressure of oxygen being attained in the oxide-filled cavity. This assumption was based on the experiments carried out under similar conditions with a carbon steel, when decarburisation of the steel occurred in the vicinity of the oxide-filled cavity. That meant that the gaseous product of decarburisation must have had a means of escape from the cavity, or decarburisation would soon have ceased. The explanation of the much larger type of globule probably lay in the greater availability of oxygen atoms in the case of oxidation in air than that from the scale. Possibly, the tendency for the formation of globules and spots during oxidation in air was, as a whole, greater, but the conversion of the steel into actual scale occurred simultaneously with the further precipitation of globules and spots in the steel.

The illustrations on Plate VI. were all of specimens in the unetched condition, with the exception of Fig. 5. Globules and spots occurred both at the grain boundaries and within the crystal grains.

[1] K. G. Lewis and U. R. Evans, *Third Report of the Corrosion Committee, Iron and Steel Institute*, 1935, pp. 173, 177.

THE PIOBERT EFFECT IN IRON AND SOFT STEEL.*

By ERIC W. FELL, Dr.-Ing., M.Sc., F.I.C.
(Manchester University and St. John's College, Cambridge).

Abstract.

When a highly polished bar of soft steel is subjected to increasing tensile stress, a value of the stress is reached at which the bar may suddenly yield to a remarkable amount with permanent surface distortion easily recognised as characteristic. The distortion is called the Piobert effect. The effect seems a property of an aggregate of iron crystals and is ordinarily not observed with other steels, non-ferrous metals and non-metallic solids. (Part I.)

The spread of the effect is compared with the yielding process. Its spread throughout the bar is progressive and occurs at nearly constant stress. It may thus be compared with the spread of solidification at constant temperature in a mass of pure metal about the melting point. A phase change is considered. (Part II.)

The distortion, analysed with ruled testbars, may be represented as successive slip of layers or planes of particles over one another. Such a layer comprises many crystals. A distorted domain is divisible into smaller domains in each of which slip may be one of two kinds. In the simpler kind, the slip seems constant in magnitude and direction. In the other kind, the magnitude of the slip in a certain direction varies. The resultant slip is generally not parallel to the surface of the rectangular testbar. The change in the shape of the testbar, after removal of the stress, may be regarded as the sum of the changes in the smaller domains. The distortion results from shear. Propagation of the effect is probably due to stress concentration. There is some similarity to distortion in a single crystal. (Part III.)

If a highly polished bar of iron or soft steel is loaded in tension, a value of the stress is reached at which the bar may suddenly yield to a remarkable amount, and this is seen to be accompanied by characteristic permanent surface distortion. The distortion figures have been named "lines" of Piobert, Lüders and Hartmann, "strain figures," "ligne ou zone de cession,"[1] "Kraftwirkungsfiguren," "Fliessfiguren," and "stretcher strains,"[2] &c. The distortion is partially shown by flaking-off of surface scale dislodged by the distortion of the metal and, on sections, by "Fry etch-figures."[3] These names all mean the same type of distortion, and confusion occurs, for it has been said that Lüders' lines do not refer to the yield point.[4] Also, "strain figures" is obscure as it may mean the "contractile cross"

* Received, May 7, 1935.

effect[5] which is quite different. The name " wedge of distortion " was used by Turner and Jevons.[6]

The distortion is really of the nature of an " effect " and as it is three-dimensional, it cannot be truly described as " lines." It was described by Piobert, Morin and Didion[7] for distortion in wrought iron caused by a projectile. Frémont[8] named the distortion after Piobert, described later by Lüders[9] and afterwards by Hartmann.[10] Piobert was thus the first to publish a description, so the name " Piobert effect " is used in this paper for the above distortion in iron and soft steel.

Part I.—Occurrence of the Piobert Effect.

Doubt exists as to whether the effect is observed other than in iron and soft steel, therefore the next pages summarise as much relevant data as are obtainable for the purpose of removing uncertainty. Perhaps the effect is observed mostly in soft steel by its frequency in engineering and also because it has a suitable surface scale, enabling the effect to be detected, as in deformed boiler-plate.

Brinell indentations made on polished surfaces of single iron crystals, kindly given by Professor G. I. Taylor, did not reveal the effect and suggest that it may be a property of an aggregate of crystals. Presence of the effect may determine the shape of the stress-strain curve, as a marked horizontal part is shown for wrought iron and soft steel and not for single iron crystals. The horizontal part is rare for other metals and solids. It seems likely that the effect would occur in very pure iron provided its crystal size be small enough. Two high-carbon steels and a stainless iron indented by a hard ball did not show the effect.[11] Seigle permanently deformed bars of soft steel after hardening, but the effect was generally absent. It was absent in deformed soft steel bars which were coarsely crystalline. It was present in 50–55 kg. steel and in soft steel with 1 per cent. of nickel, being less pronounced in 70–80 kg. steels. For soft steel case-hardened and deformed, it is not mentioned.[12]

Hartmann studied distortion effects in metals, but may not have distinguished between the Piobert and other effects, as for example the " contractile cross." Regarding non-ferrous metals, his illustrations of copper tensile bars apparently show the effect[13] (*loc. cit.* Figs. 9 and 12). Fig. 9 shows two intersecting bands on the surface, one inclined at about 58° 30′ and the other at 61° to the applied stress. The distortion in a tensile bar of German-silver is shown (Fig. 10, *ibid.*) as two intersecting bands, one inclined at about 63° and the other at 70° 30′ to the applied stress, which suggests the " contractile cross." The effect may be present in a tensile bar of delta-metal, but the illustration is not

clear (Fig. 11, *ibid.*) and the inclination of the bands is less than for the copper.

More data seem available for aluminium. Hartmann noted a wave form at constant inclination to the applied stress rapidly propagated from end to end of a bar (p. 99, *ibid.*). Hannover[11] (*loc. cit.* p. 124) reported the Piobert effect in aluminium bars of certain dimensions, but Rawdon,[14] comparing soft steel with sheet aluminium alloy, found differences in the spreading of the distortion. Anderson reports "Lüders' lines" in a duralumin tensile bar,[15] but there is no resemblance, as a wave form is shown. Portevin and Le Chatelier[16] record the effect on flat tensile bars of tempered aluminium alloy, but give no illustrations. Kuntze[17] noted two wave forms on tensile bars of polycrystalline aluminium alloy; parallel bands are propagated over the gauge length at about constant stress, and on continuing the test a second set of mutually parallel bands intersects the former set at slightly higher constant stress; the horizontal part of the stress-strain curve is in two parts. Elam[18] shows a set of parallel slip-bands on a single crystal, and of somewhat similar appearance, for a tensile bar of copper with 5 per cent. of aluminium, but, as in the previous case, the effect is absent.

An illustration is given by Nádai[19] as showing "Lüders' lines" in work-hardened copper, but the effect is absent, as there is much difference in form from the distortion in soft steel.

The effect is easily obtained on polished soft steel by Brinell indentation, so the method was used to find if the effect was obtained for other metals. It was not found in copper, zinc, nickel, aluminium, tin, cadmium, brass and bronze[11] (*loc cit.* p. 105), confirming experiments of Nádai[10] with indentations of a cylindrical stamp on hard and soft copper, cast zinc and soft iron.

It has been suggested that the Piobert effect occurs in non-metallic solids. Nádai used paraffin wax blocks to show distortion and illustrated shearing in the wax as in marble, stone or concrete (Plates XII. to XV., *ibid.*). Tests were made by the author to find if the distortion of the wax resembled the effect. Blocks with smooth bright surfaces were slowly distorted by a ball, and rapidly by bullets. The effect was absent and the distortion caused by the bullets resembled the radial and concentric cracks obtained for brittle materials such as glass, glue and pitch. It was absent in similar tests on plasticine and soap. Mr. J. B. Nevitt, of Imperial Chemical Industries, Ltd., noted distortion, like that caused by bullets fired into thin soft steel[11] (*loc. cit.* Fig. 13, Plate XII.), on the surface of common yellow bar soap in tests of hollow point bullets, though the markings were not quite so near together and only clearly defined in a band between $\frac{1}{2}$ in. and $1\frac{1}{2}$ in. from the edge of the bullet hole (private com-

munication). Osmond, Frémont and Cartaud describe distortion in stearin and gelatin due to shear,[21] but all their figures suggest that the effect is absent. The effect is macroscopical and thus the name " lignes de Lüders microscopiques " (p. 37, *ibid.*) seems inadmissible. Howe[22] gives an illustration described as " Lüders' lines " in compressed wood, but the effect is absent and shear failure of the wood is shown.

Summary.

(1) The Piobert effect commonly occurs in iron and soft steel.
(2) It is probably a property of an aggregate of crystals.
(3) It seems generally absent in non-ferrous metals.
(4) It is absent in non-metallic solids.

The effect is attributed to the plastic properties of an aggregate of iron crystals, whose mechanism of distortion may be different from that of other metals.[23] As the effect was not obtained with single crystals, it seems inadmissible directly to deduce it to be a property of the body-centred cubic lattice structure of iron.[4] An aggregate is isotropic ; the directive properties of the individual crystal are neutralised by the others present, thus the aggregate distorts more according to the general stress distribution. The effect determines the shape of the stress-strain curve. As the carbon in iron increases, the effect is suppressed, apparently due to resistance to the distortion of the ferrite crystals by the pearlite.

In non-ferrous metals, crystal size and plastic properties may be unfavourable and the only clear evidence that the effect may occur is that for copper and delta-metal.[13] The wave forms in aluminium and its alloys are different, though they may be a modification of the effect.

Part II.—The Piobert Effect and the " Yielding " Process. Change of State.

If a bar of soft steel is loaded in tension, a value of the stress is reached at which the bar may suddenly extend to a remarkable amount with the fall of the beam of the testing machine. It is called " yielding " of the bar ; meanwhile, the Piobert effect is spreading in the bar at almost constant stress. This is well known. The next experiments which were undertaken, examined especially the nature of the spreading in relation to the stress-strain curve.

Tensile bars were made from hot-rolled basic open-hearth strip, 2 in. wide by $\frac{1}{4}$ in. thick, of the following analysis : Carbon 0·06, manganese 0·36, silicon 0·01, sulphur 0·039, phosphorus 0·010, nickel 0·28, chromium 0·012 per cent. The gauge length of the testbar was 5 in., and 1 in. wide by $\frac{1}{4}$ in. thick ; the shoulders were 2 in. wide and the total length of the testbar was 12 in. They were loaded in a Buckton machine with wedge-block grips.

The balance weight on the beam was moved during a test in one direction only. An autographic curve was obtained for each bar (see Fig. 8). Results are given in Table I.

The effect was seen by the flaking-off of the surface scale, but to make clearer the extent to which it had spread certain

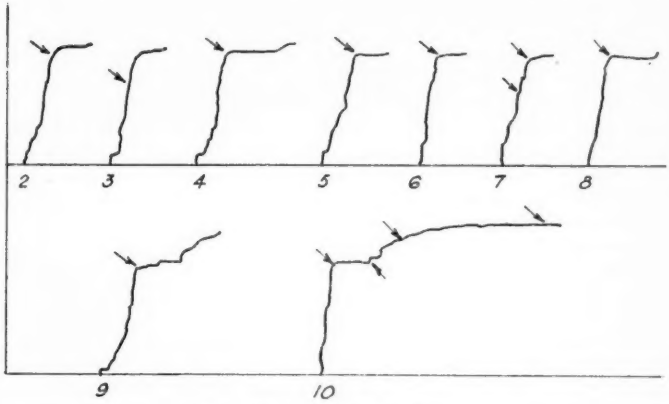


FIG. 8.—Load-Extension Curves.

TABLE I.—*Results of Experiments on the Piobert Effect.*

Test-bar No.	Stress at Appearance of Piobert Effect, Observed by Flaking-off of Scale. Tons per Sq. In.	Stress at Start of Large Horizontal Part of Curve. Tons per Sq. In.	Final Stress. Tons per Sq. In.	Extension after Removal of Load. %	External Work done in Propagating the Piobert Effect throughout the Gauge Length. Ergs per c.c. of steel.
					$\times 10^8$
2	17·0	18·7	19·1	5·5	1·59
3	13·7	18·0	18·0	4·0	1·11
4	18·7	19·1	20·9	9·0	2·22
5	17·4	17·4	17·4	4·0	1·08
6	17·1	17·1	17·1	3·7	0·98
7	10·5	17·0	17·0	3·6	0·94
8	17·2	17·2	17·6	5·6	1·49
9	16·6	17·8	21·6	10·0	1·51
10	17·6*	17·6	23·1 (max. stress)	30·0 (reduction of area at fracture = 68·8%)	1·36

* No scale, surface of bar polished.

bars were etched by Fry's method.* When the end of the horizontal part of the curves was reached most of the scale had flaked from the gauge length. In testbar No. 7, the effect was seen at a shoulder for a low stress, probably owing to stress concentration ; this is indicated by the first arrow in Fig. 8. The second arrow shows when general spread of the effect in the gauge length occurred. In testbar No. 8, the curve shows a final slight rise. The curve for No. 9 shows a " stepped " horizontal part, indicating that two slight increases of stress were necessary for spreading the effect throughout the gauge length. The stress was then increased to 21·6 tons per sq. in., with a little flaking-off of the scale in the last sloping part of the curve, due presumably to removal of tenacious remains. Testbar No. 10 had one face polished. The effect appeared suddenly at the first arrow on the curve ; its spread along the polished gauge length was complete at the second arrow. The third arrow indicates fluctuations by pressure variations in the hydraulic mains of the machine, which at the start of the curves are due to slip of the bar in the grips. The fourth arrow shows where the " contractile cross " effect was seen a few seconds before fracture. Traces of the Piobert effect remained on the bar except near the fracture where distortion was more severe. Bars Nos. 2, 3, 5, 6 and 7 showed that the scale had not all flaked off within the gauge length, and the curves show that for these the horizontal part is relatively short. No. 8 seemed more free from scale, and in Nos. 4 and 9 nearly all the scale had flaked off.

Bars Nos. 2, 5, 4 and 8 are chosen as typical (see Figs. 1 to 4, Plates VII. and VIII.). In Nos. 2 and 5 less scale had flaked off and a small amount of undistorted metal was visible in the gauge length on etching, while Nos. 4 and 8 showed more distortion. In No. 4 it would appear that the small increase of stress above that for the horizontal part of the curve was necessary to remove the last traces of undistorted metal.

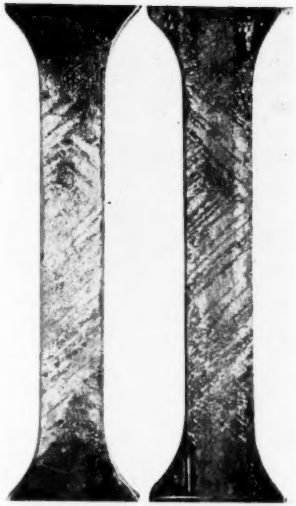
Summary.

(1) The Piobert effect generally appears suddenly at the beginning of the horizontal part of the stress-strain curve.

(2) The effect has spread throughout most of the gauge length at the end of the horizontal part, but more stress may then be necessary to cause distortion in the interstices between the already distorted metal.

The effect usually appears at a shoulder of a bar through stress concentration, and spreads towards the other shoulder. Meanwhile it may appear at the other shoulder and spread to meet the first disturbance. The initial spreading was not so great in Nos. 2, 3 and 7, for the curves show a rounded " step." The

*The surface layer was removed before etching by machining.

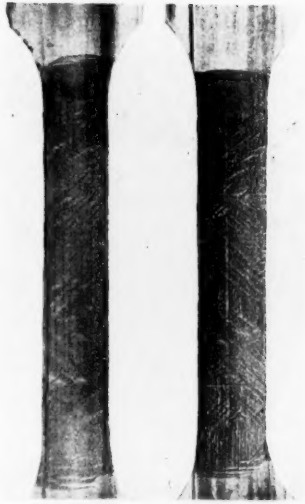


No. 2. No. 5.

FIG. 1.—Testbars, after flaking-off of surface scale, Piobert effect. × 0·42.

No. 4. No. 8.

FIG. 2.—Testbars, after flaking-off of surface scale Piobert effect. × 0·42.



No. 2. No. 5.

FIG. 3.—Etched Testbars, showing amount of spreading of Piobert effect. × 0·42.

[Ed. To face p. 80.

No. 4. No. 8.

FIG. 4.—Etched Testbars, showing amount of spreading of Piobert effect. × 0·42.

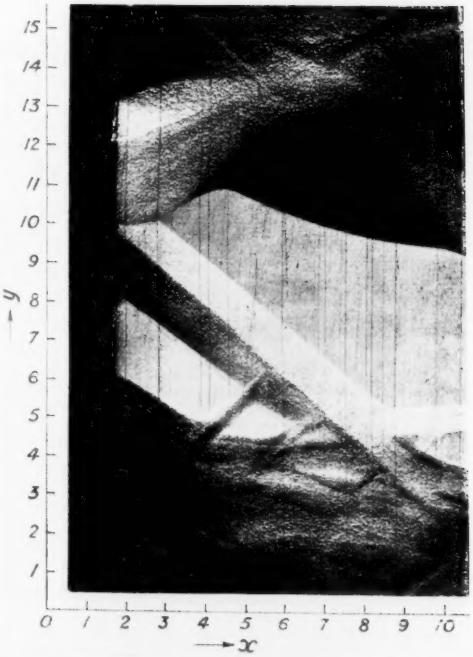


FIG. 5.—Tensile Bar *A*. Distortion of polished and ruled surface, Piobert effect. × 3·62.

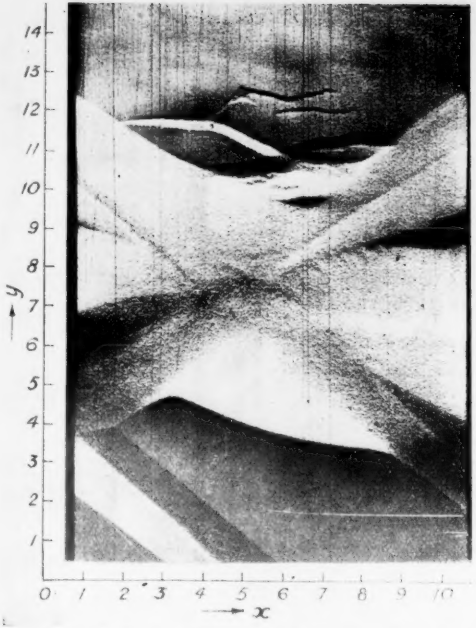


FIG. 6.—Tensile Bar *A*. Distortion of polished and ruled surface, Piobert effect. × 3·34.

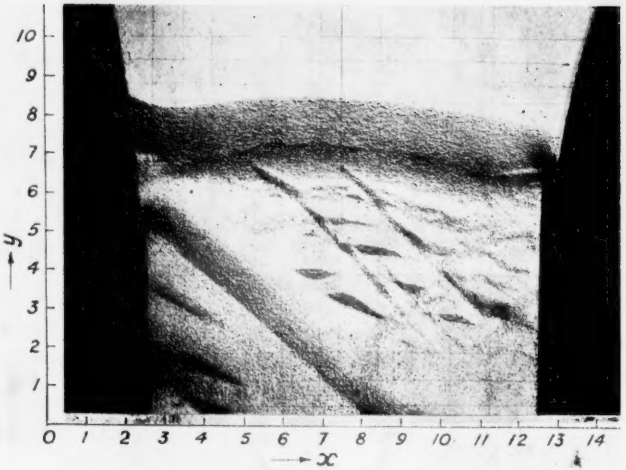


FIG. 7.—Tensile Bar *B*, shoulder. Distortion of polished surface, ruled longitudinally and transversely, Piobert effect. × 3·32.

[Feat.

greater the rate of spreading the greater the rate of extension of the bar.

A Possible Change of State.

Appearance of the effect may not necessarily mark the elastic limit in its neighbourhood, since permanent slip in the crystals may have already occurred. Spreading of the effect may occur at nearly constant stress, for in bar No. 9 two slight increases of stress were required to finish the horizontal part. Continuous increase of stress is then necessary to strain the gauge length further. Rawdon subjected bars to a series of repeated and successive loadings, each of which was just sufficient to cause yielding, and noted that spreading of the effect re-started where it had left off in the previous loading. Further distortion, where the effect was, only occurred when the effect had spread throughout the gauge length. It seemed then to be uniformly distributed[14] (*loc. cit.* p. 481). A load-time relation for bar No. 10 is given in Fig. 9.

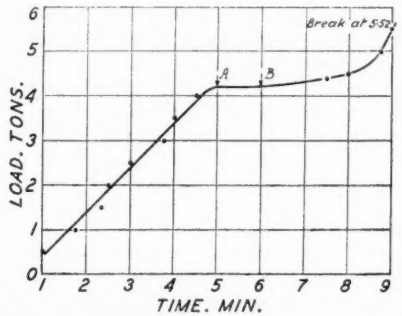


Fig. 9.—Load-Time Relation for Testbar No. 10. Arrow *A* shows when the Piobert effect was first seen on the polished bar, and arrow *B* when it completely covered the gauge length (0·239 sq. in. in original cross-section).

If an attempt be made to balance the load during yielding of a bar an irregular part is obtained through which an average horizontal may be drawn. The horizontal parts of the above curves suggest those in heating (or cooling) curves of pure metals on melting, and, for example, "load" may be substituted for "temperature." Spread of the Piobert effect throughout the metal at approximately constant stress may be compared with the spread of solidification throughout molten metal at constant temperature. The rates of spreading or of solidification vary. Undercooling near solidification may be compared with the small increase of stress, which may be necessary to start general

spreading of the effect throughout the gauge length, between the so-called upper and lower yield points. The conclusions of Körber and Pomp[24] suggest that the metal may be unstable before yielding, as the rate of loading and type of machine appear to influence the value of the upper rather than the lower point at which permanent extension has already begun. Carus-Wilson[25] compared the curves with stress-strain curves of carbon dioxide. Charbonnier and Galey-Aché[26] appear to have noted that the part of the curve between the upper and lower points represents false equilibrium, analogous to surfusion and supersaturation. This part apparently decreases as the temperature at which the steel is loaded is raised, and between 300° and 400° C. a smooth curve is obtained[19] (*loc. cit.* p. 96, Fig. 69). Instability of the solid would decrease presumably on increase of temperature. Hartmann[13] (*loc. cit.* p. 287), without giving reasons, described the analogy of Carus-Wilson as a little hazarded.

The peculiar shape of the curve for iron may be associated with a change of state. It would not be simply reversible. The external work done in spreading the Piobert effect throughout the gauge length is shown in Table I., and was found thus : Let the start and end of the horizontal part of the curves correspond to strains e_1 and e_2, respectively, and the stress at an intermediate strain be $f(e)$ where $f(e_1) = f(e_2)$; $f(e)$ is unknown but is assumed constant and equal to $f(e_1)$. The work is then the product of the stress and (e_2-e_1). In bars Nos. 4, 9 and 10 allowance was made for the increase of stress at the end of the horizontal part by means of the curves of Fig. 8 ; in Nos. 3, 6 and 7 it is possible that the horizontal part could be longer. The average of the nine values for the external work done was $1{\cdot}36 \times 10^8$ ergs per c.c. of the steel. This may vary with the pearlite present. In work by Jevons,[27] all the gauge length showed the effect at about 4 per cent. extension for open-hearth hot-rolled bars of analysis : Carbon 0·10, manganese 0·81, silicon 0·07, sulphur 0·044, phosphorus 0·03 per cent., but the work done is hardly calculable as the stress data are unknown (private communication).

A change of state generally involves change of properties and internal energy. Farren and Taylor[28] estimated the latter in plastically deformed steel containing carbon 0·17, manganese 0·76 per cent., from the external work done and the heat evolved. The " yield " gave about 2 per cent. of permanent extension. It was found for extensions of 3·99 per cent. and above that the work done was mostly converted into heat, and in nearly constant proportion. Interpolation of these results so as to include only the " yield " may not be admissible, for additional stress was imposed to produce the extension of 3·99 per cent. Below 2 per cent. the effect presumably took time to spread throughout the metal at about constant stress, and above this value distortion

occurs, presumably simultaneously everywhere in the distorted metal, at increasing stress.

The Piobert effect hardens the metal. The author[11] (*loc. cit.* Fig. 18, Plate XIV.) found that the metal was 6·8 per cent. harder than the closely adjacent undistorted metal after etching by Fry's method. Using different methods, Rawdon[14] (*loc. cit.* p. 474) found 3 to 8 per cent. increase in hardness. Moser,[29] using steel with less carbon, obtained about 14·1 to 18·8 per cent. increase, as estimated by the author. Fowler,[30] using steel of intermediate carbon content to that of the author and Rawdon, found about 4 and 8 per cent. increase. The mode of determining hardness is important. A band of distorted metal may be crossed by another band and variation of hardness may occur where they cross. The aggregate of crystals is assumed to be isotropic before distortion, and as displacement occurs in certain directions only, it may result in the aggregate becoming anisotropic, and thus the hardness may vary with direction.

X-ray studies[11] (*loc. cit.* Fig. 17, Plate XIII.) suggested that the crystals in the distorted parts were distorted more than those in the so-called undistorted parts. X-ray examination by Rawdon[14] (*loc. cit.* p. 476) failed to reveal any significant difference. Corrosion of the distorted domains, as on etching, suggests an alteration of the properties of the metal. Corrosion might be greater where two bands of distorted metal cross each other. Some spatial interpenetration of the effect may thus occur.

Such considerations suggest that the Piobert effect may correspond to a definite state of plastic strain.

Part III.—Experiments with Ruled Testbars.

The distortion associated with the Piobert effect has apparently escaped experimental study. It is assumed to be the slip of layers of the metal one over another. Previous experiment is restricted to observing the distribution of the effect in variously shaped and stressed bodies. Experiments by Gulliver[31] and Mason[32] compared the orientation of the distorted domains with stream lines and external stresses respectively.

According to the present experiments, earlier assumptions on the nature of the distortion need some revision. Hartmann assumed that the deviation of a straight ruling on a tensile-bar became 90° on entering a band of distorted metal[10] (*loc. cit.* Fig. 5, line *Cnd*), but this is far too big. A similar ruling became curved in passing through two bands, and near its middle it passes for a very short distance through undistorted metal, being deviated about 12° on entering each band[10] (*loc. cit.* Fig. 5, line *pq*), but the angle is again too big and the ruling should not be curved. It is very doubtful if two large bands can intersect as simply as shown. The shape of the ends of part of a bar which has

contracted in width may need modifying[10] (*loc. cit.*, Fig. 4, *m* and *m*), and the effect differs from the distortion at necking of a bar[10] (*loc. cit.*, Fig. 6). Frémont[33] regards this permanent contraction of the width of soft steel bars as made up of a plastically deformed part together with two elastically (yet permanently) deformed parts. The latter adjoin the ends of the former part, being the transition from it to the two undeformed ends of the bar. Permanent elastic deformation may exist adjoining such plastically deformed metal yet would not appear to make up the contraction of the bar as described, for permanent contraction appears to be formed by plastic distortion alone, as slip of layers. The shape of the boundaries of a distorted domain in a rectangular parallelepiped subjected to compression is given by Nádai.[34] Assuming slip is uniform from one face of the body to the opposite face, then the boundaries of the distorted domain should be parallel. This is not indicated, as the left-hand boundary appears as a plane and the right-hand boundary as two mutually inclined planes. Moser[29] (*loc. cit.*, p. 1603, Figs. 8 and 9) sketches the distortion in tensile bars, but it is very doubtful if the inclined bands, in which presumably uniform slip has occurred, are so uniformly orientated along the whole gauge length. The distorted layer immediately adjoining the undistorted domain is also shown as if moving over the undistorted domain to a considerable amount, but actually the larger moves to an almost infinitesimal amount only, for the rulings remain unbroken on testbars.

The present experiments on the nature of the distortion were as follows. Testbars were made from the steel strip described in Part II., with a test portion 3 in. long; the shoulders were 2 in. wide and the bar had a total length of 8 in. The test portions were polished as for microscopical work, then annealed, after which a little more polishing was needed on finer emery papers.

Lines were ruled on the polished surfaces by sliding the bars against a fixed straight-edge, with a pivoted razor blade resting on them. The blade edge was inclined about 45° to the surfaces. To inscribe uniform lines a tiny piece from the worn cutting edge was broken off at intervals. The pressure of the blade could be varied. The distortion was measured by the change in position of the rulings on photographs of the bars, using a straight-edge and protractor. The rulings are easily distinguished from chance scratches. The distortion was not simple, and to describe it without interfering with the photographs needed rectangular axes of reference. These were chosen so that the width of the bar corresponds to the x axis, the length of the bar to the y axis, and the positive direction of the z axis is taken as into the paper of the subsequent figures. The x and y axes given in the figures serve to map desired parts.

To avoid optical distortion, certain measurements were made with the ruling to be measured in the centre of the microscope field. This was impossible for the figures as they covered nearly the whole field. It is assumed that the optical error is less than the error in ruling or measurement, as the straightness and parallelism of the rulings were well maintained in the larger undistorted domains, for example, in the top of Fig. 7 (Plate IX.).

Two testbars, *A* and *B*, were treated as above.

Bar A.—Longitudinal lines only were ruled on the bar. It was loaded steadily, and at a stress of 4·5 tons per sq. in. the effect was traced at both edges of the test portion, as in the bending of a bar. The load was removed on reaching 15 tons per sq. in. The test portion was initially 0·72 in. wide by 0·215 in. thick. The distortion is described by referring to the figures.

Fig. 5 (Plate VIII.): The surface shown was near the middle of the test portion. Part of the distortion is seen as a conspicuous inclined band (the amount and direction of slip are given below). In the Figure, there are four isolated undistorted domains recognisable by a smoother surface. Their boundaries are easily traced. Their four centre points have co-ordinates ($x = 7·5$, $y = 8$), (3·2, 6·3), (5·4, 5) and (10·3, 4·3). The distortion of the rulings, once straight, is seen at the left-hand side of the Figure which shows an edge of the bar, or better, by placing a ruler against the lines.

On this left-hand side, the following two lines are typical. With the x and y co-ordinates first and second respectively in the brackets and describing the two lines as they pass down the Figure, they may be taken to start at two points (2·9, 13·6) and (4·0, 13·9). First, both lines form straight segments 12 and 7 mm. long, respectively (actually 7·2 and 4·2 mm. in the reduced Figure). There is then a change in slope of the surface of the distorted domain, where the lines deviate 1° 30′ and 1°, respectively, towards the y axis. The next segments are straight, 24 and 21 mm. long, respectively (actually 14·4 and 12·6 mm. in the reduced Figure). The former meets the join of two distorted domains in the shape of bands inclined to the y axis. The latter meets the boundary of the distorted and an undistorted domain, where it deviates 1° 30′ away from the y axis and then passes over the latter domain to the point (3·85, 9·3) on the edge of the conspicuous inclined band.

At the upper boundary of the conspicuous inclined band the lines deviate 2° 30′ and 1° 30′, respectively, from the y axis, forming straight and parallel segments 11·5 and 12 mm. long (actually 6·9 and 7·2 mm. in the reduced Figure). The band is seen to consist of two parts, a uniformly shaded upper part and a lower part shaded darker. The shading could be varied

by rotation of the object on the microscope stage; thus the distortion in the light part varied from that in the darker part, so the deviation of the lines was more carefully measured at the centre of the microscope screen (magnification $\times$ 33·3) to avoid optical distortion. Thus at the points (3·9, 8) and (4·6, 7·5) on the join of the light and darker parts, deviations of 46′ and 34′ occurred respectively, away from the y axis. The two lines are straight in the light part, being possibly very slightly curved in the darker part. The parallelism of the other lines in this band may be noted. Leaving it at (2·8, 7·4) and (3·9, 6·6), the two lines deviate each about 1° 30′ to the y axis and are parallel in the undistorted domain as straight segments 19 and 15 mm. long, respectively (11·4 and 9 mm. in the reduced Figure). Passing then to the complex domain below, they deviate from the y axis.

The distorted edge of the bar is parallel to the segments of the rulings nearby. The bar is seen to be bulged out. The two larger undistorted domains have moved relatively, as identical rulings on them are actually translated about 0·08 mm. There is possibly relative rotation of the undistorted domains on the right-hand side of the Figure. This may produce elastic stresses in them.

The other edge of the bar is just off the right-hand side of the Figure. The next measurements suggest flow towards the central longitudinal axis. Consider the lines through the points $x = 4{\cdot}6,\ 6,\ 7{\cdot}5,\ 8,\ 8{\cdot}2,\ 8{\cdot}5,\ 8{\cdot}8,\ 10{\cdot}4,\ y = 8$: Passing from the distorted domain into the large undistorted domain below, the first and second lines deviate 45′ and 30′, respectively, from the y axis, but the others deviate, in order, 7′, 30′, 30′, 45′, 45′ and 60′ towards the y axis. The angles are larger as the edge of the bar is approached.

The slip in the conspicuous inclined band was found from its component in the plane of the Figure given by the displacement of the rulings, and its component in the z direction obtained from the difference in level of two chosen points on the surface of the band focused in a microscope with calibrated fine-adjustment. By a series of readings, the join of the light and darker parts is seen to be about 0·064 mm. below the edges of the band. The two parts may be regarded as planes inclined 2° 18′ and 2° 3′ respectively to that of the Figure. The left-hand edge of the bar had also been polished, and with the inclination of the band on it, the slip was deduced parallel to the plane

$$3{\cdot}33x + 3{\cdot}97y - 2{\cdot}11z - 50 = 0. \quad . \quad . \quad . \quad (1).$$

Its amount and direction were found from the relative displacement of the two points (4·2, 9) and (3·5, 8·2). Assuming the first to be at rest, the resultant relative slip of the other point is 0·102 mm. The resultant slip makes angles of 53° 6′, 120° 12′, 51° with the

positive direction of the x, y and z axes, respectively. Considering the accuracy of the measurements, the angle of 13° 47′ which the direction of the resultant slip makes with the plane (1) is not a serious error. The upper and light part of the band is 1·56 mm. wide. The relative resultant displacement of its sides was 0·102 mm. The number of crystals was 41 per mm.; thus each crystal contributed to a displacement of about one-fourteenth of its width. The component of the contribution in a direction parallel to the length of the bar, and expressed as a percentage of the width of the crystal, is 3·5 per cent., and is somewhat less than the extensions of those testbars when the Piobert effect had spread through the entire gauge length (Part II.). The shear force, considered as acting parallel to the slip plane so that plastic distortion may occur, was about $1{\cdot}01 \times 10^9$ dynes per sq. cm. The shear strain, considered as the ratio of the relative lateral movement of two parallel planes of slip to the distance between them, was 0·07. The stress normal to these planes was $1{\cdot}5 \times 10^9$ dynes per sq. cm.

Some interpenetration of two distorted domains may occur about the point (5·2, 5·9). Presumably, this happens only to a small extent since the general spread of distortion occurs at nearly constant stress.

Fig. 6 (Plate IX.): The surface shown was near the middle of the test portion and adjoins that of the previous Figure. Both edges of the bar are shown, and similar axes are attached. The cruciform shape of the large distorted domain is of interest; here minute necking of the bar occurred, and the left- and right-hand edges contracted 0·14 and 0·091 mm., respectively. It suggests the "contractile cross" effect occurring at much greater strains near fracture. Closer study shows the sides of the bar to consist of straight segments parallel to the rulings nearby. Thus the distortion of the metal is of a discontinuous nature compared to the "contractile cross" effect, where the distorted surfaces are smoothly curved. The causes of both effects may be similar. The surface discontinuities in the domain appear straight, tending to meet at (4·8, 7·2) near the centre of the domain. This suggests flow in their direction. The top of the Figure and its bottom right-hand corner show undistorted domains.

The distortion was not very complex, and a description of the three rulings through the points (2·9, 10), (4·55, 10) and (8·7, 10) is typical. The first ruling leaves the undistorted domain at the point (2·9, 11·2) and deviates 1° from the y axis; it then passes to (2·95, 8·4) where it deviates 45′ towards the y axis. At (2·95, 7·3) it deviates about 1°, and 1° at (2·9, 6·5), both towards the y axis. Entering the undistorted domain at (2·9, 4·6) it deviates about 1° 30′ from the y axis. The second ruling, on entering the distorted domain, devia es 30′ from the y axis. After passing

about half-way through, it deviates 30′ towards the y axis. Entering the undistorted domain below, it deviates 30′ from the y axis. The third ruling has more segments, with deviations comparable to the above but generally in a correspondingly opposite direction. The domain may thus be divided into presumably plane areas containing straight segments, which are not generally parallel in an area, since deviations decrease as the longitudinal central axis of the bar is approached. It suggests that the slip decreases where such wedge-shaped areas exist.

The roughly triangular light area with corners at (2·5, 4·5), (8·8, 3·3) and (5·2, 6·5) appears to have been formed later than the two inclined bands adjoining, and may serve to equalise stresses near the centre of the bar caused by the flow in the bands. The same tendency to interaction of two inclined bands is seen at the top of the Figure. The rulings are straight in the triangular area, but on the right- and left-hand sides the deviation is towards the central axis.

Bar B.—Longitudinal and transverse lines were ruled on the bar. The test portion was initially 0·715 in. wide by 0·22 in. thick. It was loaded steadily to 16·8 tons per sq. in., then the load was removed. The effect spread quickly over the surface of the bar.

Fig. 7 : Distortion at a shoulder of the testbar is shown. There were small accidental notches in the shoulder about the points (2·6, 5·3) and (12·8, 6·5). A broad and regular area of distortion at the shoulder crosses from one edge of the bar to the other. The longitudinal rulings deviate in passing from the shoulder into this area. The deviations of the six rulings on the right-hand of the figure, in order from the edge, were 45′, 20′, 10′, 10′, $<10'$ and 5′, and on the left-hand side, in order from the edge, were 60′, 45′, 45′, 45′, $>0'$, $>0'$, towards the central line. The rulings in the area are very nearly straight and suggest flow towards the central line. The only transverse ruling clearly traceable through the area is that at $y = 8{\cdot}4$; it is straight from one edge of the bar to the other. Below the area the distortion is more complex.

Part of the bar adjoining that shown in Fig. 7 was examined. Bands of distorted metal extended from one edge of the bar to the other edge, inclined at about − 131° to the y axis. The transverse lines in an undistorted domain deviated, on an average, about 1° 15′ from the x axis on entering a band. The longitudinal rulings deviated, on an average, about 1° 38′ towards the y axis in passing upwards and entering this band. In a similar way as before, it was deduced that slip in the band is parallel to the plane $3{\cdot}65x + 4{\cdot}05y - 50 = 0$ within a certain region. The variation in the width of the bar caused by slip is shown in Fig. 10. The bar was divided longitudinally into $\frac{1}{2}$ in. lengths whose change

on distortion δ, is plotted. Its width is plotted also. The abscissa is the distance along the bar from shoulder to shoulder. The test portion was not all distorted. The two curves vary inversely. The part of the bar shown in Fig. 7 corresponds to the abscissa interval 3·8 to 3·25, and the other shoulder to 0 to 0·5. Width

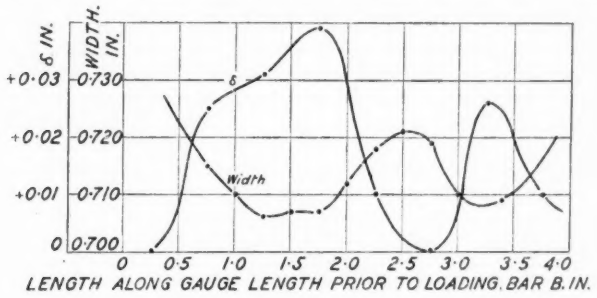


FIG. 10.

variation along testbars during the yielding of silicon constructional steel is given by Kuntze and Sachs[35]; the usual horizontal part of the stress-strain curve may consist of two horizontal parts.

Summary and Discussion.

The purpose of Part III. of this paper is to describe experiments on the nature of the distortion *within* the distorted domains associated with the Piobert effect. Differences (discussed on pp. 83 and 84) were found between the distortion visualised by other workers and that obtained in the experiments, but the usual assumption that the distortion is slip of layers of particles over each other is confirmed.

A distorted domain may be divided into smaller domains or areas in which the slip may be represented by a model consisting of a pack of cards capable of sliding over one another, the cards being ribbed or grooved so that they can slide over each other in one direction. In the simplest case, it seems that the slip may be represented by the sliding of each card a constant amount over its neighbour, since the rulings remain straight and parallel (*cf.* the conspicuous inclined band in Fig. 5, and p. 85). In less simple cases, the rulings remain straight but not parallel, suggesting variation of the amount of sliding (*cf.* the broad area of distortion at the top of Fig. 7, and p. 88).

The cause of the slip is shear, as in a part of the gauge length of a tensile bar away from the shoulders, where stress concentration may be expected, there is a marked tendency for the distorted domains to form prominent bands inclined to the directions of the applied stress at angles near those of maximum shear stress.

The slip velocity of the layers is reduced to zero—presumably by work-hardening. The growth of distorted domains suggests that the areas in which slip is uniform are not formed by instantaneous slip of all layers but rather by successive slip. There is evidence[36] that the traces of slip planes in individual crystals are nearly parallel in the distorted domains.

Since the Piobert effect resembles the slip of planes of particles, it may be like the distortion of the single crystal, and its mechanism may also be similar. Taylor considered materials which lose their power to withstand shear stresses when they yield, and the yielding of successive grains might be expected to produce an increasingly heterogeneous distribution of internal stresses. To trace the effect of this heterogeneity on the yield properties of the whole granular mass, a hypothetical material was imagined with the property that when any portion yields all shear stresses vanish but the compressibility remains unaltered. This material had the property that the greatest stress concentrations occur when elongated faults lie at 45° to the directions of the principal stresses, so that faults, once started, would propagate themselves along lines at 45° to these directions. This property seems to have its counterpart in real materials which can exhibit the Piobert effect.[37]

An aggregate of iron crystals seems, then, to favour this uniform propagation of faults. Propagation of the Piobert effect is probably due to stress concentration. The nature of the effect seems such as to bring the aggregate into a condition where this particular propagation of faults does not occur on increase of stress.

The author is greatly obliged to Messrs. Vickers-Armstrongs, Limited, Naval Construction Works, Barrow-in-Furness, and in particular to Sir C. W. Craven, for kindly granting facilities for the experiments. Also, he expresses his thanks for help to Mr. M. C. Oldham, F.C.S., and Mr. A. Casley, of their engineering laboratories.

The steel was kindly given by The Barrow Haematite Steel Company, Limited.

The advice of Professor F. C. Thompson, D.Met., of Manchester University, is gratefully acknowledged.

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(7) *Mémorial de l'Artillerie*, 1842, vol. 5, p. 505.
(8) *Génie Civil*, 1919, vol. 74, p. 51 (footnote).
(9) *Dingler's Polytechnisches Journal*, 1860, vol. 155, p. 18.
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(30) *Engineering*, 1931, vol. 132, pp. 299 and 420.
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(33) *Génie Civil*, 1923, vol. 82, p. 177, Fig. 9.
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CORRESPONDENCE.

Dr. Ing. A. NÁDAI (East Pittsburgh, Pa., U.S.A.) wrote that the regular pattern of lines, which could be observed on the surface of pieces of mild steel just after the stresses had reached the plastic limit, had been studied by engineers who had reported on these strain lines a number of times. Although many points with reference to the laws of the formation of these regular patterns had been fully described, other details were still perhaps not quite sufficiently known and were worth while being investigated. The efforts of the author of the present paper to investigate carefully the formation of strain figures on deformed steel therefore deserved to be commended.

The paper by Dr. Fell left on the writer the impression that, according to the author's belief, the formation of these regular patterns was characteristic particularly or exclusively of mild steel. The spirals around the impression of a cylindrical stamp left on the polished surface of a mild-steel block mentioned in the paper would be an example of the effect. When a round stamp or a hard steel ball was pressed into mild steel the typical logarithmical spirals would be seen on the surface, whereas if the same test was repeated with soft copper or aluminium, spirals could not be observed, only a uniform area of flow. That had been confirmed both by the author of the paper and by the writer in experiments made several years ago. It seemed, however, that from the negative result in the case of a concentrated pressure test made with copper or aluminium it did not conversely follow that the effect was completely absent in such metals as copper or aluminium. The formation of thin layers of plastic material could be influenced artificially. If the conditions were made similar to those which were thought to be essential for the formation of strain figures in steel, localisation of flow could be found. The author quoted a number of observations describing surface patterns of a regular nature obtained with other materials. On soft paraffin-wax cylinders compressed in an axial direction very regular helical lines crossing each other had been seen. When repeating the experiment mentioned above with the stamp, however, spiral lines again did not appear on paraffin wax. H. Meyer and F. Nehl[1] had demonstrated by etching an axial plane cross-section of a compression cylinder of mild steel that conical surfaces of slip formed during a compression test, starting at the edges. This same localised effect of deformation (namely, conical surfaces of shear) had, on the other hand, been found in compressed paraffin-wax cylinders by the writer. That, in the case of the experiment with a

[1] *Stahl und Eisen*, 1925, vol. 45, p. 1961.

stamp on paraffin-wax, spiral lines did not become visible was due to the extremely low tensile strength of this material. Where tensile stresses were present, uniform yielding was prevented by the formation of tension cracks. It was, however, noteworthy that, as in the case of the internal conical surfaces of slip just mentioned, in simple compression tests, in axial sections through a paraffin-wax block, after an impression of a stamp had been made, a curved line could be detected in the interior, starting from the edges of the impression. Strain lines with a similar orientation had been found in the interior of blocks under concentrated pressure areas in mild steel.

Surfaces of slip had also been studied in perfectly loose materials, such as sand. It might perhaps appear at a first glance strange to connect here observations of such a heterogeneous nature as the formation of strain lines in a plastic metal, such as mild steel, with that of the surfaces of slip in perfectly loose materials, such as sand. There was, however, an important reason why this had an advantage : It was given by the fact that the orientation of all these surfaces accompanying yielding or partial or total fracture was so intimately and regularly related to the orientation of the principal axes of stress in deformed

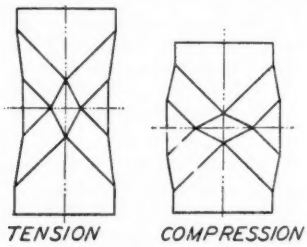

FIG. A.

bodies. It had been variously pointed out that certain geometrical properties and the laws of formation of these surfaces of slip were essentially the same or closely related. Their presence in plastically-deformed metals might be inferred by analogy or by observation. (The surfaces of slip had a definite relation in their orientation to the state of strain, and the orientation of the resulting stream lines of plastic flow again depended on them.)

From a mechanical point of view, it was highly interesting to note the great difference between the effect of a small disturbance or of a slight local concentration of stress upon the corresponding disturbance produced in the state of strain in elastic bodies and in the plastic state of deformation. In elastic bodies the increase of strain was limited to the small region of stress concentration.

When materials yielded under a sharply defined yield stress, however, and this stress was just reached, the effect was a completely different one. A localised disturbance of stress now caused a disturbance of strain which spread immediately to very large distances (compared with the dimensions of the region where the stresses were higher). Plastic strains would appear in two thin layers, starting from the point of stress concentration. This was illustrated by the sketch, Fig. A. The surfaces along which disturbances or discontinuities in the distribution of strains travelled were the surfaces of slip.[1]

Another point of a metallurgical nature perhaps deserved to be mentioned. A number of observations would seem to indicate to the writer that in mild steel the ferrite grains were surrounded by a rigid and brittle skeleton or framework. At the upper yield point this skeleton, which had prevented plastic slipping of the ferrite grains under lower stresses, collapsed. As soon as it was sufficiently weakened and broken, the grains could start to deform plastically, in the same way as a single iron crystal would do. In soft copper there was no solid frame preventing the first slipping of the crystal grains.

With regard to Fig. 8, the writer observed that in these load/extension curves no indication was given of the upper and lower yield points, which were a characteristic feature of the formation of strain figures in tensile tests with mild steel. This was due probably to the type of testing machine used in the experiments. The author mentioned that during the tests the balance weight on the beam was moved only in one direction. This prevented the observation also of the upper yield point, which was believed to be closely associated with the first starting of a strain figure.

Summing up the preceding, it seemed to the writer that the phenomenon of macroscopical slip, which was so impressively demonstrated through the formation of regular strain figures in mild steel and the recent observations of the author of the present paper, was perhaps a general and essential feature of the mechanism of plastic flow in an aggregate of crystals, solidly joined, or of the first movement from a state of equilibrium in quite loose materials. The recent advances made in finding a comparatively simple mechanism for the discontinuous slip in single metallic crystals (G. I. Taylor's " dislocations " travelling through the

[1] They were the analogue to the surfaces along which, for example, a discontinuity in pressure would travel in a compressible liquid. These surfaces (or curves in plane problems) were known as the " characteristics " of the corresponding differential equations. In the case of plastic strain it could be shown that the surfaces of slip were the characteristics of the differential equations for the strain or displacement components. Such surfaces, along which discontinuities would spread, were impossible in elastic bodies ; they were, however, a consequence of the particular type of differential equations obtained in the case of plastic bodies.

lattice) would seem to indicate that microscopical slip in a crystal as well as the macroscopical slip in aggregates of crystals, in which thousands of grains were moved at once in a number of distinct layers (mild steel), were related mechanical phenomena. Similarly, as the thermal agitation in the lattice, together with the presence of some irregularity in the binding of the elements of a lattice structure, caused a discontinuous effect and slip occurred first in individual planes instead of in a continuous way, the plastic equilibrium in mild steel temporarily (just when the plastic deformations started) became unstable. For continuous yielding throughout the whole region of highest stresses, a slightly higher resultant force was needed than that necessary for maintaining non-uniform yielding in thin layers. In the cases when there was no definite yield stress and plastic limit (soft copper), the conditions at first were different, and yielding, on the contrary, must affect the whole aggregate gradually. At higher stresses (strain-hardened copper), however, cases of partial yielding in thin layers were still possible when a slight stress concentration or a soft inclusion localised the plastic flow.

Mr. HENRY S. RAWDON (Department of Commerce, Bureau of Standards, Washington, U.S.A.) wrote that, in his opinion, the practical importance of studies of the " Piobert effect " was not fully realised, as a rule, by materials-testing engineers. In particular, the fact that yielding in soft steel was a decidedly localised phenomenon, for which considerable time was required for it to progress throughout the entire volume of metal under stress, was not appreciated. Too often, it was tacitly assumed in conducting a tension test that the " permanent set " which might remain on releasing the load after yielding had occurred, was uniformly distributed throughout the reduced section of the specimen. The prolonged discussions and controversy concerning the value and significance of " yield point " and " elastic limit " determinations in tension testing would undoubtedly have been greatly reduced had such discussions been predicated upon accurate and adequate knowledge of the deformation changes in steels which studies of the Piobert effect served to demonstrate so strikingly.

The newer concept of " yield strength "—that was, the stress required to produce a specified permanent set, for example, 0·002 in. per in.—was now favoured by most American materials-testing engineers. That was a distinct advance over the older vague concepts of " yield point " and " elastic limit." It was, however, decidedly empirical in its nature and might not fully " tie in " with the facts. Studies of the strain markings on mild-steel specimens under tension, corresponding to the yielding which started at and progressed inwardly from the ends of the

reduced section of the bar under load, carried out by the writer several years ago and referred to by the author, had shown that for long specimens (18 in.) the " yield strength " of the material as calculated from the total permanent set and the nominal gauge length could be reached before the greater part of the specimen had been permanently altered at all.

In some materials, however, of which the austenitic steels were notable examples, yielding in a tension bar under stress was distributed very uniformly throughout the entire reduced portion of the specimen. It followed, then, that in the latter case, yield strength—defined, for example, as the stress producing a permanent set of 0·002 in. per in.—had a definite physical meaning, whereas in the other case it might be an entirely fictitious quantity.

Another useful application which the writer had made of the Piobert effect was for testing the accuracy of the initial axial alignment of flat specimens in the tension tests. Almost invariably, the initial " yield " resulted in the formation of a " wedge " of deformed metal near the fillet at each end of the bar. These two wedges were located, one on each side of the bar. Their formation was the first indication of yielding, and their position was indicative of lack of absolutely true axial alignment of the specimen. That had been observed repeatedly, even with the best " self-aligning " grips obtainable.

AUTHOR'S REPLY.

The AUTHOR, in reply, wrote that Dr. Nádai's contribution was much appreciated. The Piobert effect and the " contractile cross " effect in soft steel, the patterns in copper and the wave forms in aluminium, the spiral and other patterns in wax, &c., were regarded as fundamentally related because of their regular nature and orientation relative to the direction of the applied stress and of the maximum shear stress. Since different solids varied in plastic behaviour under load, the patterns might be expected to vary. In particular, those in soft steel, which were given the name of " Piobert effect," were regarded as a group by themselves, owing to their distinctive form, associated as they were with the sudden and extensive yield peculiar (in general) to iron and soft steel. The paper was meant to be a study of the Piobert effect ; it was an interesting property of an aggregate of ductile crystals (iron) and it demonstrated very simply slip along planes through the aggregate.

In Brinell tests described in the paper, copper, aluminium and other metals (excluding iron and soft steel) did not show the Piobert effect, and the only conclusion which it was desired to

draw from this was that the samples would not show the effect if they were tested differently, as in tension for example. Regarding Fig. 8, it was true that little indication was given of the so-called lower yield point, since the balance weight (connected to the load recorder) on the beam of the machine was moved in one direction only during a test. The so-called upper yield point was regarded as the point at which the large yield began.

Mr. Rawdon's contribution was very welcome. His view as to the value of studies of the effect was appreciated, for not only were they concerned with the behaviour of aggregates of crystals, but with the strength of materials as well.

THE PROPERTIES OF SOME LOW-NICKEL STEELS CONTAINING MANGANESE.*

By R. H. GREAVES, D.Sc. (Research Department, Woolwich).

Abstract.

The mechanical properties of steels containing 0·3 to 0·4 per cent. carbon, 0·7 to 1·35 per cent. manganese, and 0·5 to 2·0 per cent. nickel have been examined.

In the normalised condition a steel containing 0·35 per cent. carbon, 1 per cent. manganese and 1 to 1·5 per cent. nickel gives better mechanical properties than those previously obtained from 3 per cent. nickel steel, 1·4 per cent. manganese steel and silico-manganese high elastic limit steels.

Steels of approximately this composition do not harden fully when quenched in oil in sections ¾ in. thick, and are subject to mass effect, but their mechanical properties in the oil-hardened and tempered condition are equal to those for which 3 and 4 per cent. nickel steels of similar carbon and low manganese contents are frequently employed.

It would appear that, in suitable circumstances where a yield point of over 32 tons per sq. in. is required, with maximum load over 45 tons per sq. in. and impact figure over 40 ft.-lb., oil-hardened and tempered steels containing 0·35 to 0·40 per cent. of carbon and 1·0 to 1·3 per cent. of manganese with about 1 per cent. of nickel may be used with economic advantage in sections up to 2½ in. in thickness in place of 3 per cent. nickel steel. If a high impact figure is required and the section to be treated exceeds 2½ in. in thickness, it may be necessary to reduce the carbon to 0·30 and increase the nickel to 1·5 or even 2·0 per cent., since steels with the higher carbon content indicated above, when treated in large sections, give an impact figure below 40 ft.-lb.

Apart from the investigations of Guillet,[1] and of Carpenter, Hadfield and Longmuir,[2] very little published information about manganese-nickel steels is available. Jones[3] investigated the influence of manganese on nickel steels, but did not increase the manganese content above 0·8 per cent. nor reduce the nickel content below 3 per cent. The mechanical properties of some manganese-nickel-copper steels are quoted by Rhodes.[4]

It has been shown by Jones[5] that in the normalised condition a steel containing 1·4 per cent. of manganese has mechanical properties not unlike those of a steel of similar carbon content containing 3 per cent. of nickel, and the present investigation was undertaken to ascertain whether a part of the nickel in nickel steels could be replaced by manganese. The steels examined contained up to 2 per cent. of nickel and up to 1·35 per cent. of manganese. They can be divided conveniently into four groups:

* Communication from the Research Department, Woolwich, received June 1935.

Group.	Steels.[1]	Carbon. %	Manganese. %	Nickel. %
I.	1 to 4	0·3	1·1	0·5 to 2·0
II.	6 to 9	0·3	1·35	0·5 to 2·0
III.	10 to 13	0·4	0·7	0·5 to 2·0
IV.	14 to 17	0·4	1·0	0·5 to 2·0
	5	0·35	1·2	1·0

Normalised Steels.

Specimens of the steels were normalised at a series of temperatures between 860° and 740° C., and their mechanical properties determined. In all cases the best properties were obtained by normalising at temperatures in the neighbourhood of 800° C.

TABLE I.—*Tensile, Brinell Hardness and Notched-Bar Impact Tests (Longitudinal) on Steels Normalised and Slowly Cooled from* 800° C.

Sections : 5 × 1½ × ¾ in.

Steel No.	Carbon. %	Manganese. %	Nickel. %	Treatment.*	Elastic Limit.[1] Tons per Sq. In.	Yield Point.[1] Tons per Sq. In.	Maximum Load. Tons per Sq. In.	Yield Ratio.	Elongation. %	Reduction of Area. %	Average Brinell Hardness Number.	Average Izod Impact Value. Ft.-Lb.
1	0·34	1·04	0·45	N.	27	28·0	41·6	0·67	31	62	190	40
				S.C.	21	24·0	39·7	0·60	27	55	176	38
2	0·33	0·99	0·97	N.	25	26·6	41·9	0·63	30	64	192	36
				S.C.	22	24·0	39·9	0·60	29	55	176	26
3	0·33	1·04	1·45	N.	29	30·6	44·0	0·70	31	64	202	53
				S.C.	27	28·4	42·0	0·68	29	57	188	44
4	0·32	1·08	1·95	N.	23	31·0	46·4	0·67	27	58	211	53
				S.C.	26	29·4	44·1	0·67	28	57	197	41
5	0·36	1·18	1·03	N.	30	32·4	45·9	0·70	30	64	213	66
				S.C.	25	27·4	43·9	0·62	29	63	200	49
6	0·32	1·35	0·47	N.	26	28·2	43·4	0·65	31	68	197	43
				S.C.	24	25·6	41·2	0·62	29	57	183	38
7	0·33	1·35	1·03	N.	28	31·4	47·0	0·67	28	64	216	50
				S.C.	28	30·0	44·6	0·67	28	59	198	34
8	0·30	1·32	1·49	N.	14	28·0	47·4	0·59	26	56	219	27
				S.C.	24	27·0	43·7	0·60	28	57	194	41
9	0·29	1·27	2·06	N.	10	27·0	50·8	0·53	23	43	233	21
				S.C.	17	25·8	45·1	0·57	26	55	199	29
10	0·40	0·66	0·45	N.	24	26·0	41·0	0·63	30	56	183	31
				S.C.	21	22·0	39·2	0·58	29	51	170	22
11	0·36	0·71	1·02	N.	25	27·0	41·7	0·65	29	57	190	32
				S.C.	22	24·4	39·9	0·61	28	52	174	25
12	0·40	0·68	1·47	N.	27	29·0	43·9	0·66	27	55	200	30
				S.C.	23	25·8	42·2	0·61	26	49	182	24
13	0·41	0·69	1·98	N.	27	29·2	45·4	0·65	27	56	207	29
				S.C.	23	26·0	43·6	0·60	26	50	192	21
14	0·38	1·10	0·43	N.	27	28·6	45·1	0·63	28	61	202	33
				S.C.	24	26·0	42·6	0·61	28	55	187	25
15	0·39	1·10	1·00	N.	29	30·8	47·8	0·64	27	59	220	38
				S.C.	24	26·8	45·4	0·59	27	55	201	27
16	0·38	1·07	1·37	N.	28	30·8	47·1	0·65	27	59	215	37
				S.C.	25	26·8	45·4	0·59	27	56	199	23
17	0·37	1·03	1·96	N.	20	30·0	47·8	0·63	26	55	221	31
				S.C.	26	28·4	45·9	0·62	26	58	203	24

*N = Normalised. S.C. = Slowly cooled at 14° C. per min.

[1]See note on p. 113.

The mechanical properties of the steels in the longitudinal direction after normalising at 800° C., and also after cooling at a rate of about 14° C. per minute are given in Table I. The effect of nickel on the mechanical properties of the normalised steels is shown in Fig. 1.

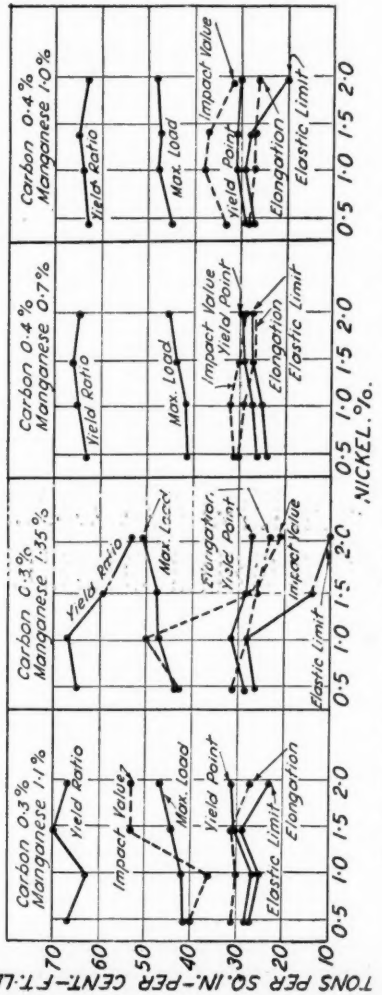

FIG. 1.—Steels Normalised at 800° C. Effect of nickel on the mechanical properties.

Addition of both manganese and nickel to the steels improves the mechanical properties without serious loss of ductility, but increase of the total alloy content beyond certain limits causes a tendency to air-hardening. The limits of manganese and nickel content which will avoid this tendency appear from the present work, and that of Jones[5] to be :—

Manganese. %	Nickel. %
0·7	3·0
1·0	1·5
1·35	1·0

Steels 4, 8, 9 and 17, in which these limits are exceeded, all show the wide interval between elastic limit and yield point in the normalised condition which is characteristic of steels which tend to be air-hardened. The effect is most marked in steels 8 and 9, and tempering these steels after normalising shows definitely that there has been some air-hardening (see Table II.).

TABLE II.—*Tensile, Brinell Hardness and Notched-Bar Impact Tests on Steels Normalised at* 800° C. *and Tempered for* 2 *hr. followed by Cooling in Air.*

Sections : 5 × 1¼ × ⅜ in.

Steel No.	Carbon. %	Manganese. %	Nickel. %	Tempering Temperature. °C.	Elastic Limit. Tons per Sq. In.	Yield Point. Tons per Sq. In.	Maximum Load. Tons per Sq. In.	Yield Ratio.	Elongation. %	Reduction of Area. %	Average Brinell Hardness Number.	Average Izod Impact Value. Ft.-Lb.
8	0·30	1·32	1·49	N.T.*	14	28·0	47·4	0·59	27	56	219	27
				400	23	29·6	47·1	0·63	29	59	219	38
				500	26	30·6	45·9	0·67	29	58	216	46
				600	29	30·8	43·4	0·71	31	64	203	56
9	0·29	1·27	2·06	N.T.*	10	27·0	50·8	0·53	23	43	233	21
				400	22	31·0	49·3	0·63	25	57	231	23
				500	23	30·6	46·4	0·66	26	57	216	41
				600	27	31·6	44·0	0·72	28	63	209	49

* N.T. = Not tempered.

A comparison of the properties of steels 3 and 5 with the best of the manganese, nickel and silico-manganese high elastic limit steels examined by Jones[5] (see Table III.) shows the nickel-manganese steels to be superior in every respect. The 1·4 per cent. manganese steel of Jones is approaching the limit beyond which the mechanical properties can no longer be improved by addition of manganese alone, since a manganese content in excess

TABLE III.—*Comparison of Mechanical Properties of Normalised Nickel, Manganese, Silico-Manganese and Nickel-Manganese Steels.*

Steel No.	Carbon. %	Manganese. %	Silicon. %	Nickel. %	Normalising Temperature. °C.	Elastic Limit. Tons per Sq. In.	Yield Point. Tons per Sq. In	Maximum Load. Tons per Sq. In.	Yield Ratio.	Elongation. %	Reduction of Area. %	Average Brinell Hardness Number.	Average Izod Impact Value. Ft.-Lb.
*A**	0·39	0·60	..	2·98	800	26	28·9	42·1	0·69	29	61	183	48
B	0·38	1·39	..	..	800	24	27·9	44·4	0·63	28	58	197	31
C	0·32	1·31	0·93	..	860	26	29·8	45·6	0·65	30	63	200	44
3	0·33	1·04	..	1·45	800	29	30·6	44·0	0·70	31	64	202	53
5	0·36	1·18	..	1·03	800	30	32·4	45·9	0·70	30	64	213	66

* *A, B, C*: Steels examined by Jones[9]

of 1·6 per cent. makes the steel liable to air-harden, while 3 per cent. of nickel seems to be about the useful limit for nickel structural steels.

It appears, therefore, that the combined use of nickel and manganese will give mechanical properties better than can be obtained by the use of either alloying element singly.

Steels 1 to 9 were tested in the transverse direction after normalising and after slow cooling at 14° C. per minute from 800° C. The results, given in Table IV., show good ductility in this direction.

The microstructures of the normalised steels showed a uniformly small grain size of which Fig. 4 (Plate X.) (steel 3) may be taken as typical. The tendency of some of the steels to air-harden is shown in Fig. 5 (steel 9).

Oil-Hardened and Tempered Steels.

The mechanical properties of the steels in the oil-hardened and tempered condition are given in Table V.

Variation of the hardening temperature over the range 820° to 1,000° C. does not appear to affect the mechanical properties greatly, except in the cases of steels 10, 11, 12 and 14. A high hardening temperature causes a marked reduction of the impact value in these four steels. Since the first three of these steels are in the lower manganese series, and steel 14 contains only 0·5 per cent. nickel, it appears that both manganese and nickel raise the upper limit of temperature from which hardening can be carried out without the mechanical properties being adversely affected.

Although most of the steels do not suffer a serious deterioration of properties by hardening from so high a temperature

TABLE IV.—*Tensile, Brinell Hardness and Notched-Bar Impact Tests (Transverse) on Steels Normalised and Slowly Cooled from* 800° C.

Sections : 5 × 1¾ × ¾ in.

Steel No.	Carbon. %	Manganese. %	Nickel. %	Treatment.*	Elastic Limit. Tons per Sq. In.	Yield Point. Tons per Sq. In.	Maximum Load. Tons per Sq. In.	Yield Ratio.	Elongation. %	Reduction of Area. %	Average Brinell Hardness Number.	Average Izod Impact Value. Ft. Lb.
1	0·34	1·04	0·45	N.	22	25·4	42·2	0·60	27	54	188	30
				S.C.	21	24·0	40·4	0·59	30	52	180	27
2	0·33	0·99	0·97	N.	21	24·8	42·4	0·58	28	56	193	29
				S.C.	19	22·6	40·4	0·56	29	54	180	22
3	0·33	1·04	1·45	N.	28	31·0	44·8	0·69	27	55	202	40
				S.C.	23	27·6	42·6	0·65	27	57	192	33
4	0·32	1·08	1·95	N.	20	29·2	47·4	0·62	26	51	212	32
				S.C.	25	28·4	44·5	0·64	26	55	199	29
5	0·36	1·18	1·03	N.	.	29·2	46·3	0·63	28	57	207	40
				S.C.	25	27·6	44·1	0·63	27	52	197	32
6	0·32	1·35	0·47	N.	22	27·2	44·2	0·62	26	54	198	34
				S.C.	23	25·8	42·0	0·61	27	50	186	27
7	0·33	1·35	1·03	N.	24	31·4	47·8	0·66	25	50	216	31
				S.C.	26	30·4	45·5	0·67	26	54	203	26
8	0·30	1·32	1·49	N.	11	26·2	49·5	0·53	22	43	220	22
				S.C.	22	26·4	44·5	0·59	26	52	202	27
9	0·29	1·27	2·06	N.	9	26·8	51·9	0·52	21	39	234	19
				S.C.	16	26·4	45·4	0·58	25	50	204	27

* N. = Normalised. S.C. = Slowly cooled at 14° C. per min.

as 1,000° C. there seems to be a slight advantage in using a lower hardening temperature, probably about 850° C.

Fig. 2 shows the mechanical properties of the steels oil-hardened from 900° C. and tempered at 600° C. Both manganese and nickel improve the elastic properties, maximum load and yield ratio while causing only a slight reduction in ductility.

Although the steels were not fully hardened by quenching in oil, the properties after hardening and tempering compare favourably with those of 3 and 4 per cent. nickel steels treated in the same section,[3] and it seems likely that in certain cases some of the steels could be used in place of steels with a considerably higher nickel content.

Mechanical Properties of Steels Treated in Various Sections.

The effect of mass on the mechanical properties of a selection of the steels oil-hardened from 900° C., tempered at 600° C. and cooled in air is shown in Table VI.

The steels containing 0·4 per cent. carbon all showed a considerable reduction in impact value when the size was increased beyond 1¾ × ¾ in., but this effect did not appear in the steels containing 0·3 per cent. carbon up to a cross-section of 3¼ × 2¼ in.

TABLE V.—*Tensile, Brinell Hardness and Notched-Bar Impact Tests on Steels Oil-Hardened and Tempered for 2 hr., followed by Cooling in Air.*

Sections: 5 × 1¼ × ¾ in.

Steel No.	Carbon. %	Manganese. %	Nickel. %	Hardening Temperature. °C.	Tempering Temperature. °C.	Elastic Limit. Tons per Sq. In.	Yield Point. Tons per Sq. In.	Maximum Load. Tons per Sq. In.	Yield Ratio.	Elongation. %	Reduction of Area. %	Average Brinell Hardness Number.	Average Izod Impact Value. Ft.-Lb.
1	0·34	1·04	0·45	1000	600	28	33·6	46·2	0·73	23	63	218	55
				900	550	31	33·4	46·9	0·71	24	65	213	45
					600	26	31·0	45·1	0·69	29	66	208	60
					650	25	28·4	42·3	0·67	30	67	195	58
				820	600	28	31·0	44·8	0·69	29	66	208	66
2	0·33	0·99	0·97	1000	600	30	33·6	45·8	0·73	26	64	221	36
				900	550	29	35·6	48·7	0·73	24	61	233	32
					600	28	35·0	47·2	0·74	26	65	223	43
					650	23	30·6	43·3	0·71	28	68	199	49
				820	600	32	34·4	46·5	0·74	27	66	218	48
3	0·33	1·04	1·45	1000	600	33	41·6	51·2	0·81	22	62	234	54
				900	550	34	38·6	50·2	0·77	23	63	247	67
					600	31	36·0	47·6	0·76	25	64	228	69
					650	24	32·0	43·5	0·74	27	67	200	76
				820	600	35	37·6	47·8	0·79	27	65	220	70
4	0·32	1·08	1·95	1000	600	36	44·0	53·0	0·83	24	65	251	54
				900	550	34	42·6	53·4	0·80	22	62	262	51
					600	37	40·6	50·9	0·80	24	63	244	63
					650	34	36·0	46·2	0·78	28	66	216	70
				820	600	40	42·4	50·7	0·84	25	65	242	66
5	0·36	1·18	1·03	1000	600	36	44·8	53·6	0·84	22	57	260	52
				900	550	33	39·4	51·4	0·77	23	62	238	62
					600	27	35·6	47·7	0·74	24	64	221	64
					650	25	32·4	44·5	0·73	28	67	205	85
				820	600	34	37·0	47·6	0·78	27	65	223	69
6	0·32	1·35	0·47	1000	600	31	36·6	47·8	0·77	22	67	227	35
				900	550	31	35·4	48·2	0·73	25	62	230	38
					600	27	33·8	46·3	0·73	25	64	220	50
					650	27	31·0	43·3	0·72	25	64	195	65
				820	600	31	35·6	47·2	0·75	25	64	224	81
7	0·33	1·35	1·03	1000	600	33	40·8	51·6	0·79	23	59	247	47
				900	550	36	41·4	53·0	0·78	24	59	246	55
					600	34	38·0	49·3	0·77	27	63	233	62
					650	31	34·0	44·7	0·76	27	65	209	67
				820	600	35	38·0	49·2	0·77	25	63	234	72
8	0·30	1·32	1·49	1000	600	37	45·0	53·6	0·84	22	62	260	31
				900	550	37	43·0	53·5	0·80	22	57	252	44
					600	35	41·2	51·0	0·81	22	61	236	66
					650	29	34·4	45·2	0·76	25	62	210	57
				820	600	35	39·4	49·7	0·79	23	58	246	74

In connection with the impact values it should be mentioned that although the oil-hardening was carried out in the sizes stated, the tempering was carried out on sections $1\frac{3}{4} \times \frac{3}{4}$ in. from the centre of the larger pieces. The air-cooling after tempering was, in consequence, rather more rapid than would have been the case

TABLE V.—*Continued*

Steel No.	Carbon. %	Manganese. %	Nickel. %	Hardening Temperature. °C.	Tempering Temperature. °C	Elastic Limit. Tons per Sq. In.	Yield Point. Tons per Sq. In.	Maximum Load. Tons per Sq. In.	Yield Ratio.	Elongation. %	Reduction of Area. %	Average Brinell Hardness Number.	Average Izod Impact Value. Ft.-Lb.
9	0·29	1·27	2·06	1000	600	39	46·4	54·0	0·86	23	63	264	34
				900	550	36	46·6	54·7	0·85	22	58	270	45
					600	38	43·0	51·6	0·83	25	64	250	61
					650	31	36·0	45·7	0·79	27	69	219	84
				820	600	38	43·8	51·6	0·86	24	65	251	65
10	0·40	0·66	0·45	1000	600	22	27·6	45·5	0·61	27	64	204	10
				900	550	28	30·4	46·8	0·65	25	58	208	31
					600	27	28·8	43·9	0·66	29	62	198	39
					650	26	27·0	41·0	0·66	31	64	181	36
				820	600	24	28·6	44·0	0·65	32	65	205	39
11	0·36	0·71	1·02	1000	600	20	27·8	45·5	0·61	25	57	203	6
				900	550	27	31·0	46·0	0·67	26	58	201	34
					600	27	28·2	43·1	0·65	27	65	193	44
					650	24	27·0	41·1	0·66	31	66	182	41
				820	600	28	30·0	43·4	0·69	29	64	198	46
12	0·40	0·68	1·47	1000	600	30	36·4	48·1	0·76	24	66	224	18
				900	550	30	32·8	49·1	0·67	24	61	220	25
					600	29	30·6	45·6	0·67	29	64	207	35
					650	25	28·6	43·0	0·67	30	65	197	36
				820	600	25	30·4	45·6	0·67	30	64	206	37
13	0·41	0·69	1·98	1000	600	31	36·0	47·8	0·75	25	66	221	47
				900	550	36	39·0	51·9	0·75	21	62	241	46
					600	32	33·0	47·4	0·70	26	65	222	47
					650	24	30·4	44·0	0·69	29	68	203	51
				820	600	32	35·0	47·6	0·73	28	66	216	49
14	0·38	1·10	0·43	1000	600	24	28·4	46·2	0·61	25	63	209	4
				900	550	27	31·8	48·6	0·65	26	62	217	29
					600	28	30·0	45·1	0·67	30	66	207	36
					650	27	28·6	43·1	0·66	30	68	193	42
				820	600	26	29·6	44·9	0·66	30	66	198	41
15	0·39	1·10	1·00	1000	600	32	40·8	51·1	0·80	23	65	234	69
				900	550	32	37·0	50·4	0·73	25	65	225	49
					600	33	34·4	47·2	0·73	26	65	215	68
					650	30	32·8	44·3	0·74	29	67	202	76
				820	600	30	33·4	46·6	0·72	28	65	208	74
16	0·38	1·07	1·37	1000	600	28	36·0	47·5	0·76	23	65	222	45
				900	550	33	39·8	52·3	0·76	23	61	247	63
					600	33	34·8	46·8	0·74	27	66	210	65
					650	30	31·2	43·8	0·71	27	66	204	71
				820	600	30	34·0	47·2	0·72	28	66	224	72
17	0·37	1·03	1·96	1000	600	37	39·4	49·1	0·80	27	65	236	70
				900	550	31	40·6	52·7	0·77	25	62	249	64
					600	32	35·8	47·2	0·76	28	66	224	78
					650	23	30·4	45·2	0·67	28	66	209	78
				820	600	33	35·8	47·2	0·76	25	66	222	70

had the full section been tempered, and in those steels susceptible to temper-brittleness (see later) the impact value obtained may be slightly high.

The elastic properties of all the steels are affected appreciably

by increase of cross-section, but the effect in the steels containing 0·3 per cent. of carbon is no greater than that found in 3 and 4 per cent. nickel steels by Jones.[3]

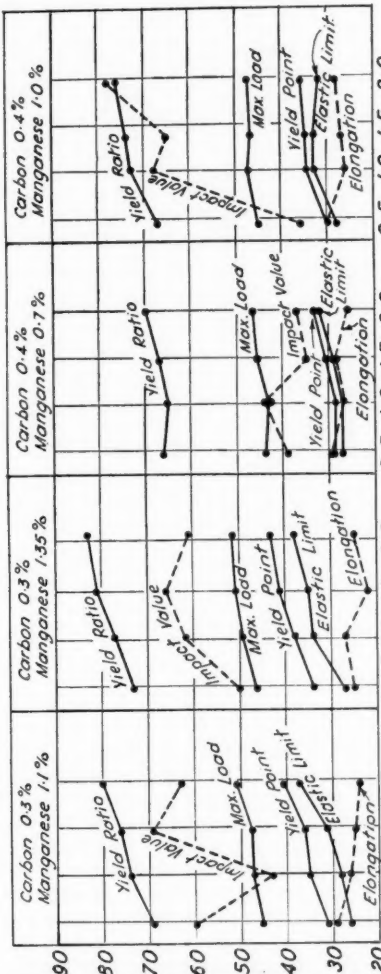

Fig. 2.—Steels O.–H. 900° C., T. 600° C., followed by air-cooling. Effect of nickel on the mechanical properties.

TABLE VI.—*Tensile, Brinell Hardness and Notched-Bar Impact Tests on Steels Treated in Different Sections. Oil-Hardened from 900° C., and Tempered for 2 hr. at 600° C., followed by Cooling in Air.*

Steel No.	Carbon. %	Manganese. %	Nickel. %	Cross-Section. In.	Elastic Limit. Tons per Sq. In.	Yield Point. Tons per Sq. In.	Maximum Load. Tons per Sq. In.	Yield Ratio.	Elongation. %	Reduction of Area. %	Average Brinell Hardness Number.	Average Izod Impact Value. Ft.-Lb.
3	0·33	1·04	1·45	1¾ × ¾	31	36·0	47·6	0·76	25	64	228	69
				1¾ × 1¼	31	35·0	47·2	0·74	26	63	226	54
				1¾ × 2¼	31	33·0	45·4	0·73	28	65	207	63
				2¼ × 3¼	30	32·0	43·2	0·74	28	64	207	62
4	0·32	1·08	1·95	1¾ × ¾	37	40·6	50·9	0·80	22	62	244	63
				1¾ × 1¼	36	39·2	49·7	0·79	24	62	235	54
				1¾ × 2¼	31	36·8	47·9	0·77	26	63	226	65
				2¼ × 3¼	32	36·4	47·6	0·76	26	63	218	62
8	0·30	1·32	1·49	1¾ × ¾	35	41·2	51·0	0·81	22	61	236	66
				1¾ × 1¼	34	40·4	50·5	0·80	23	62	243	51
				1¾ × 2¼	33	37·2	48·4	0·77	23	64	217	59
				2¼ × 3¼	30	34·0	46·1	0·74	26	64	215	59
9	0·29	1·27	2·06	1¾ × ¾	38	43·0	51·6	0·83	25	64	250	61
				1¾ × 1¼	33	40·0	49·7	0·80	22	61	243	54
				1¾ × 2¼	35	39·6	49·7	0·80	24	64	238	58
				2¼ × 3¼	33	37·6	48·3	0·78	26	64	224	61
15	0·39	1·10	1·00	1¾ × ¾	33	34·4	47·2	0·73	26	65	215	68
				1¾ × 1¼	30	33·0	48·7	0·68	26	64	219	34
				1¾ × 2¼	31	32·6	48·2	0·68	26	62	212	30
				2¼ × 3¼	30	32·8	48·4	0·68	25	62	214	32
16	0·38	1·07	1·37	1¾ × ¾	33	34·8	46·8	0·74	27	66	210	65
				1¾ × 1¼	30	33·0	47·8	0·69	26	63	217	39
				1¾ × 2¼	31	33·0	47·4	0·70	28	63	213	38
				2¼ × 3¼	30	32·2	48·0	0·67	27	61	209	36
17	0·37	1·03	1·96	1¾ × ¾	32	35·8	47·2	0·76	28	66	224	78
				1¾ × 1¼	32	34·8	48·5	0·72	25	62	225	56
				1¾ × 2¼	32	33·8	47·8	0·71	26	62	221	50
				2¼ × 3¼	27	32·4	47·8	0·68	26	61	211	43

Water-Hardened and Tempered Steels.

As stated above, the steels did not harden fully on being quenched in oil, but quenching in water from 850° C. led to full hardening of almost all the steels in a cross-section $1\frac{3}{4} \times \frac{3}{4}$ in.

Figs. 6 and 7 show the structures of steel 3 quenched in oil and in water, respectively, and tempered at 600° C.

The mechanical properties of the steels after water-hardening from 850° C. and tempering at various temperatures are given in Table VII. Fig. 3 shows the properties after tempering at 600° C.

As in the oil-hardened and tempered condition, manganese and nickel both lead to improvement of the mechanical properties without reducing the elongation greatly.

TABLE VII.—*Tensile, Brinell Hardness and Notched-Bar Impact Tests on Steels Water-Hardened from 850° C. and Tempered for 2 hr., followed by Cooling in Air.*

Sections: 5 × 1½ × ½ in.

Steel No.	Carbon. %	Manganese. %	Nickel. %	Tempering Temperature. °C.	Elastic Limit. Tons per Sq. In.	Yield Point. Tons per Sq. In.	Maximum Load. Tons per Sq. In.	Yield Ratio.	Elongation. %	Reduction of Area. %	Average Brinell Hardness Number.	Average Izod Impact Value. Ft.-Lb.
1	0·34	1·04	0·45	550	37	42·8	53·6	0·80	22	61	256	55
				600	37	39·8	50·2	0·79	22	61	247	57
				650	35	36·0	44·8	0·80	28	68	203	70
2	0·33	0·99	0·97	550	41	47·4	55·9	0·85	20	58	274	55
				600	39	42·8	51·2	0·84	23	63	248	61
				650	34	36·4	45·9	0·79	26	68	221	71
3	0·33	1·04	1·45	550	46	49·8	57·2	0·87	22	59	275	46
				600	43	45·8	52·9	0·87	23	61	256	53
				650	39	40·8	47·2	0·86	27	67	225	69
4	0·32	1·08	1·95	550	51	53·2	58·8	0·90	21	58	287	46
				600	43	47·0	53·4	0·88	23	62	263	52
				650	41	42·6	47·8	0·89	27	67	227	66
5	0·36	1·18	1·03	550	46	50·0	57·2	0·87	22	62	273	48
				600	42	46·0	52·3	0·89	24	62	252	53
				650	37	40·8	47·1	0·87	27	67	222	64
6	0·32	1·35	0·47	550	42	48·2	56·5	0·85	21	59	272	54
				600	42	44·8	52·6	0·85	23	61	256	56
				650	37	38·0	46·2	0·82	27	68	219	71
7	0·33	1·35	1·03	550	48	51·0	58·8	0·87	22	57	281	42
				600	43	45·0	53·2	0·85	23	61	239	48
				650	38	40·4	47·6	0·85	27	66	228	62
8	0·30	1·32	1·49	550	47	50·8	58·3	0·87	21	58	283	44
				600	43	46·0	53·1	0·87	22	61	254	54
				650	39	41·0	47·9	0·86	27	68	227	65
9	0·29	1·27	2·06	550	50	52·8	58·9	0·90	21	58	283	38
				600	44	46·8	53·3	0·88	22	62	256	50
				650	39	40·0	47·2	0·85	27	68	225	65
10	0·40	0·66	0·45	550	37	39·0	52·1	0·75	24	61	264	33
				600	30	33·4	48·4	0·69	25	65	231	38
				650	30	31·4	44·8	0·70	29	67	211	49
11	0·36	0·71	1·02	550	37	42·4	52·8	0·80	24	62	256	60
				600	36	40·4	49·6	0·81	25	64	237	71
				650	29	32·4	44·3	0·74	28	67	215	75
12	0·40	0·68	1·47	550	39	44·8	54·6	0·82	23	61	270	57
				600	43	45·6	52·0	0·88	23	64	249	64
				650	38	39·6	46·4	0·85	27	66	224	72
13	0·41	0·69	1·98	550	47	51·4	57·7	0·89	22	62	277	52
				600	43	45·6	53·2	0·86	24	64	248	64
				650	38	39·8	47·5	0·84	28	67	229	76
14	0·38	1·10	0·43	550	45	48·0	53·8	0·89	22	62	270	56
				600	39	42·0	51·2	0·82	25	65	249	66
				650	35	37·2	45·7	0·81	28	67	219	77
15	0·39	1·10	1·00	550	48	53·8	59·8	0·90	22	57	284	49
				600	43	48·6	55·8	0·87	23	62	264	55
				650	36	40·4	48·9	0·83	27	67	226	69
16	0·38	1·07	1·37	550	49	50·6	57·3	0·88	22	59	278	53
				600	43	45·0	52·7	0·85	25	64	250	63
				650	38	40·0	47·0	0·85	28	67	226	76
17	0·37	1·03	1·96	550	48	50·6	58·0	0·88	21	59	279	52
				600	43	45·4	53·2	0·85	23	64	255	64
				650	39	41·4	47·2	0·88	28	67	226	78

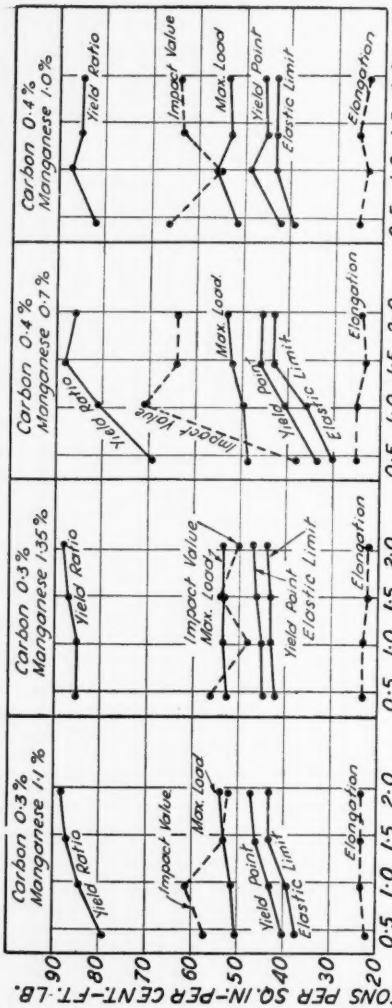

Fig. 3.—Steels W.–Q. 850° C., T. 600° C., followed by air-cooling. Effect of nickel on the mechanical properties.

The effect of nickel in improving the response of a steel to hardening treatment is clearly illustrated by comparing the portions of Figs. 2 and 3 relating to a steel containing 0·3 per cent. carbon and 1·35 per cent. manganese. A steel of this com-

position with no nickel does not harden fully on oil-quenching from 900° C., and the addition of nickel up to 2 per cent. effects a steady improvement in the mechanical properties. Water-quenching the same steel from 850° C. causes it to harden fully, and the addition of nickel up to 2 per cent. is almost entirely without effect on the properties.

Although a high manganese content is generally considered to cause liability to cracking in heat treatment, no such effect was encountered on the present series of steels. In neither the oil-quenched nor water-quenched sections were any cracks observed.

Temper-Brittleness.

Susceptibility to temper-brittleness was determined by a standard method,[6] and the results are shown in Table VIII.

TABLE VIII.—*Susceptibility to Temper-Brittleness.*

Steel No.	Carbon. %	Manganese. %	Nickel. %	Impact Value. Ft.-Lb.		Susceptibility Ratio.
				O.-H. 900° C., T. 650° C. Reheated to 600° C. 2 hr. and cooled—		
				In Water.	½° C. per min.	
1	0·34	1·04	0·45	57	67	0·85
2	0·33	0·99	0·97	47	42	1·1
3	0·33	1·04	1·45	59	63	0·9
4	0·32	1·08	1·95	57	40	1·4
6	0·32	1·35	0·47	58	46	1·3
7	0·33	1·35	1·03	51	27	1·9
8	0·30	1·32	1·49	48	20	2·4
9	0·29	1·27	2·06	50	15	3·3
10	0·40	0·66	0·45	35	32	1·1
11	0·36	0·71	1·02	40	32	1·3
12	0·40	0·68	1·47	47	35	1·3
13	0·41	0·69	1·98	79	72	1·1
14	0·38	1·10	0·43	50	37	1·4
15	0·39	1·10	1·00	82	61	1·3
16	0·38	1·07	1·37	87	34	2·6
17	0·37	1·03	1·96	79	28	2·8

Nickel alone, or in the presence of a low percentage of manganese, does not induce susceptibility to temper-brittleness,[6]

but the well-known effect of manganese in rendering a steel more susceptible is reinforced by an increase in the nickel content. The highest susceptibility ratios for steels of a given carbon content, with manganese over 1 per cent., were obtained in those containing the highest percentage of nickel.

Although some of the steels have susceptibility ratios of over 2, there is no indication in the results given in Table V. that air-cooling of small sections after tempering results in brittleness. In larger sections air-cooled after tempering the rate of cooling may become sufficiently slow to reduce the impact value.

Summary and Conclusions.

The mechanical properties of steels containing 0·3 to 0·4 per cent. carbon, 0·7 to 1·35 per cent. manganese, and 0·5 to 2·0 per cent. nickel have been examined.

The amount of nickel which may be added to a steel containing manganese without inducing a tendency to air-hardening decreases from well over 2 per cent. with a manganese content of 0·7 per cent. down to 1 per cent. when the manganese content is 1·35 per cent.

In the normalised condition a steel containing 0·35 per cent. carbon, 1 per cent. manganese and 1 to 1·5 per cent. nickel gives good mechanical properties, better than those previously obtained from 3 per cent. nickel steel, 1·4 per cent. manganese steel and silico-manganese high elastic limit steels.

Steels of approximately this composition do not harden fully when quenched in oil in sections $\frac{3}{4}$ in. thick, and are subject to mass effect, but their mechanical properties in the oil-hardened and tempered condition are equal to those for which 3 and 4 per cent. nickel steels of similar carbon and low manganese contents are frequently employed.

An increase in manganese and nickel, either separately or together, within the limits investigated, diminishes the mass effect and leads to an improvement in the mechanical properties without serious loss of ductility. It also increases the range of temperature from which quenching may be carried out without adversely affecting the mechanical properties. Steels of the higher manganese and nickel contents are susceptible to temper-brittleness, but air-cooling of sections $\frac{3}{4}$ in. thick does not give rise to a low impact figure. All the steels investigated hardened fully on water-quenching, and the relatively high manganese content of some of them did not result in any cracking.

It would appear that, in suitable circumstances where a yield point of over 32 tons per sq. in. is required, with maximum load over 45 tons per sq. in. and impact figure over 40 ft.-lb., oil-hardened and tempered steels containing 0·35 to 0·40 per cent. of carbon and 1·0 to 1·3 per cent. of manganese with about 1 per

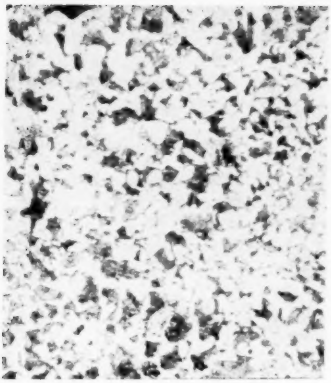

FIG. 4.—Steel No. 3, normalised at 800° C. × 250.

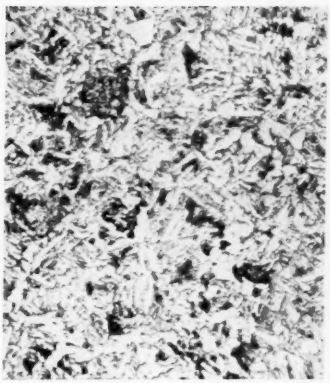

FIG. 5.—Steel No. 9, normalised at 800° C. × 250.

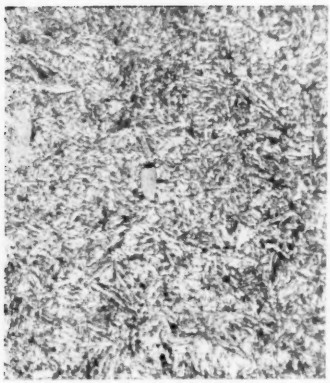

FIG. 6.—Steel No. 3, oil-hardened from 900° C., tempered at 600° C. × 250.

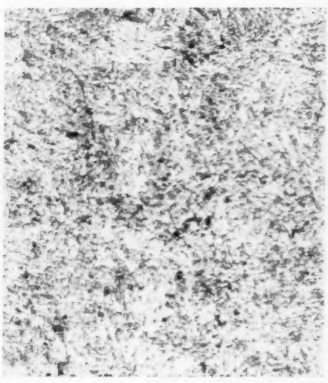

FIG. 7.—Steel No. 3, water-hardened from 850° C., tempered at 600° C. × 250.

cent. of nickel may be used with economic advantage in sections up to 2½ in. in thickness in place of 3 per cent. nickel steel. If a high impact figure is required and the section to be treated exceeds 2½ in. in thickness, it may be necessary to reduce the carbon to 0·30 and increase the nickel to 1·5 or even 2·0 per cent., since steels with the higher carbon content indicated above, when treated in large sections, give an impact figure below 40 ft.-lb.

The author desires to record his indebtedness to Mr. G. Burns, M.Met., a member of the staff of the Research Department, Woolwich, who carried out the experimental work.

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(2) H. C. H. CARPENTER, R. A. HADFIELD and P. LONGMUIR: "Seventh Report to the Alloys Research Committee," *Proceedings of the Institution of Mechanical Engineers*, 1905, p. 857.
(3) J. A. JONES: *Journal of the Iron and Steel Institute*, 1928, No. I. p. 295.
(4) J. B. RHODES: *Iron Age*, 1915, vol. 96, p. 1553.
(5) J. A. JONES: *Journal of the Iron and Steel Institute*, 1929, No. II. p. 127.
(6) R. H. GREAVES and J. A. JONES: *Journal of the Iron and Steel Institute*, 1925, No. I. p. 246.

NOTE ADDED OCTOBER, 1935.

The steels were made in a high-frequency induction furnace. The average silicon content was 0·3 per cent., the sulphur was less than 0·030 per cent. and the phosphorus less than 0·025 per cent. The elastic limit given in the tables is the limit of proportionality determined with the Ewing extensometer in the first loading of the test-piece. The yield point is the load giving a permanent extension of 0·2 per cent.

CORRESPONDENCE.

Mr. J. A. JONES (Sheffield) wrote that the paper by Dr. Greaves describing the mechanical properties of nickel steels in which part of the normal nickel content was replaced by manganese constituted a particularly valuable contribution to the literature of alloy steels.

The properties of some of the steels within the range of composition covered by the paper were fairly familiar, and most people had longed for the opportunity of investigating in a systematic manner a comprehensive series of steels such as that examined by the author. The information was now made available and it would prove of immense value to the steel supplier and steel user from the point of view of obtaining improved mechanical properties and the possibility of obtaining these at a reduced cost.

As far as the writer's own companies were concerned, they had standardised a number of nickel-manganese steels, and the ranges of analyses adopted to give optimum properties within specified ranges of maximum stress agreed, in a remarkable manner, with the recommendations made in the paper, and they were very pleased to obtain confirmation of their findings.

With regard to steel in the normalised condition, it was considered that a further saving in cost, without depreciation in mechanical properties, might be accomplished by replacing more nickel by manganese.

Raising the manganese still higher and reducing the nickel content to under 1 per cent. provided a steel, which, although showing a tendency towards air-hardening, was particularly suitable for service in the normalised and tempered condition.

Large tonnages of steel containing carbon 0·35 to 0·40 per cent., with manganese about 1·2 per cent. and nickel about 1 per cent. were supplied regularly (replacing 3 per cent. nickel oil-hardening steel) in sections up to 3 in. square to give an ultimate stress of 48 tons per sq. in. minimum.

Excellent mechanical tests were obtained in the oil-hardened and tempered condition, a selection taken at random being as follows :

Section. In.	Ultimate Stress. Tons per sq. in.	Yield Point. Tons per sq. in.	Elongation on 2 in. %	Reduction of Area. %	Average Izod Impact Figure. Ft.-lb.
$2\frac{1}{8}$ square .	49·9	39·7	24	57	67
$2\frac{1}{2}$ „ .	50·5	40·0	23	57	53
$3\frac{1}{4} \times 2\frac{1}{4}$.	51·2	40·1	22·5	57	52

The steel was also used to a considerable extent in the water-hardened and tempered condition and still improved results were

obtained without any trouble with the incidence of cracking. No appreciable drop in the impact figures occurred in sections of the dimensions under consideration if air-cooling from the tempering temperature was adopted instead of water-cooling.

Nickel-manganese steels of slightly higher nickel content were also supplied where a maximum stress of 55 tons per sq. in. was specified, in sections up to 2½ in. in diam. For such requirements, the steel constituted a particularly cheap substitute for nickel-chromium steels, although manganese-molybdenum steels produced by the companies with which the author was associated were now generally supplied to meet such tests, as it had been found that that type of steel provided, over a very wide range of tensile strength, mechanical properties superior to those of plain nickel and nickel-chromium steels at practically the same cost as the nickel-manganese steels.

Mr. C. W. HIRST (Sheffield) wrote that steels containing 1 to 1½ per cent. of manganese had been extensively used to supply the demand for steels having a tensile strength of 40 tons per sq. in. or over and impact figures higher than those which could be attained when using plain carbon steels, and the series of steels described in the paper could be regarded as a further step in the same direction.

To those who were daily engaged in the manufacture of axles, high-torque motor shafts, crusher spindles and the like from such steels, the figures presented in Fig. 1 would come somewhat as a surprise. In the case of the steel containing carbon 0·3 and manganese 1·35 per cent., an increase in the nickel content from 1 to 2 per cent. brought about a reduction not only of the yield ratio but of the actual yield point of the material.

To a lesser extent the same effect was present in the case of the 0·4 per cent. carbon, 1·0 per cent. manganese steel. The addition of 1 to 2 per cent. of nickel was often made to this steel in making large alternator rotors, where increased strength was required without any reduction in the magnetic permeability of the steel forming the rotors.

In none of the applications mentioned had the writer noted this reduction of yield point with the increase of nickel content, and, bearing in mind that the masses dealt with would be much greater than those tested by Dr. Greaves, he would like to know if the author could put forward any explanation of the variation in behaviour of these steels when different masses were being employed.

AUTHOR'S REPLY.

Dr. GREAVES, in reply, stated that he was glad to have Mr. Jones' confirmation of the results of the work done at Wool-

wich. It showed that the properties obtainable in these steels of low nickel content had not been exaggerated, but rather underestimated in the paper. Since normalising of many of these steels constituted a mild form of hardening, considerable improvement could be obtained by the normalising and tempering treatment which Mr. Jones advocated. The fall in yield point and other properties which accompanied an increase of nickel from 1 to 2 per cent. in the normalised steels (a feature to which Mr. Hirst drew attention) was due to incipient air-hardening which occurred in steels of this manganese and nickel content, cooled through the critical range at the rate employed, *viz.*, 14° per min. To restore the properties, tempering must follow normalising of material of the thickness in question ($\frac{3}{4}$ in.). In larger masses, to which Mr. Hirst referred, the natural rate of cooling in air would be slower, and air-hardening would not set in until a higher manganese nickel content was reached. The first indication of air-hardening in most steels was its detrimental effect on the elastic properties, an effect which could easily be avoided by cooling more slowly, or be overcome more satisfactorily by tempering after normalising.

The author agreed that the steels described in his paper were not likely to prove so generally useful as manganese-molybdenum steels; in fact, their properties were such that the addition of a little molybdenum was called for, thus converting them into nickel-manganese-molybdenum steels, but for certain purposes (one of which was indicated by Mr. Hirst) the nickel might still confer advantages in the presence of molybdenum.

THE USE OF THE HELE-SHAW APPARATUS IN THE INVESTIGATION OF THE FLOW OF METALS.*

By A. M. HERBERT and Professor F. C. THOMPSON.
(Department of Metallurgy, the University of Manchester.)

Summary.

The Hele-Shaw apparatus has been used by engineers in the study of stream-line flow of fluids with interesting results. So far as the authors are aware, however, no attempt has hitherto been made to use it in the investigation of the deformation of metals. Since it appears that these materials also flow in a stream-line fashion both in hot- and in cold-working, a study has here been made concerning the correspondence which may be obtained between the flow-lines obtained with the apparatus and those actually found in the metals themselves. Typical drop-forgings, tubes and particularly wire have been investigated and the degree of agreement is good. It would, therefore, appear that it is legitimate to employ this type of apparatus where the mode of flow during any process of deformation is required to be known. In the design of the dies for the production of drop-forgings, for instance, it is probable that invaluable evidence could be obtained of a kind which is not otherwise readily available.

Part I.

Ever since the publication of Tresca's classical experiments it has been generally recognised that under a sufficiently high intensity of stress, metals may flow in a manner essentially identical with that of a very viscous fluid. As is now well known, the flow of any fluid may be divided into two quite distinct categories. As the pressure is increased the flow which is initially stream-line in character is suddenly converted into one of turbulent motion. As the viscosity of the fluid is increased the pressure required for the latter type of flow is also increased rapidly, and in the case of metals the evidence is overwhelming that turbulent flow never occurs during any working process. Even in materials such as shear steel or Damascene steel where the macrostructure might, at first sight, suggest turbulent flow, there is little doubt that the irregularity is merely due to the localised nature of the deformation under the individual blows of the hammer. It follows, therefore, that determinations of the nature of the stream-line flow of any fluid should throw considerable light on that of a metal during both hot- and cold-working processes.

* Received June 21, 1935.

Apparatus.

One of the most direct, and at the same time simplest, methods of examining the nature of stream-line flow is with the aid of the apparatus first described by Hele-Shaw[1] in 1897 and modified and improved in the following year. The apparatus, which is shown in Fig. 1 and Fig. 4 (Plate XI.), consists of two glass plates *A* and *G* which are separated by a model made of drawing paper of the contour to be examined. The liquid used is glycerine, the clear fluid being introduced through the union *B*, whence it passes into the cavity *C* and then flows down within the die. Another supply of glycerine, but coloured with a red aniline dye, enters at *D*, and passes into another cavity ground in *A*, which is closed by a brass plate *E* at the lower edge of which a series of small, equidistant holes are drilled. Alternating streams of red and clear glycerine thus pass down between the plates and reveal the stream-lines characteristic of the particular shape of die under investigation. The red and clear glycerine are stored in two air-tight containers, Fig. 4, into which air under pressure is pumped, the pressure in each case being the same. Tests made regarding the influence, if any, of the pressure upon the nature of the flow in no case showed any alteration as this was increased. Under the maximum pressure available—about 70 or 80 lb. per sq. in. in the containers—the flow was invariably stream-line, but by using a fluid of lower viscosity, such as water, and a much thicker die it is possible to obtain turbulent flow. So far, however, as the present work is concerned, the pressure employed was about 10 lb. per sq. in. A permanent record of the results may be obtained by placing the plates, strongly clamped together, in a vertical container with openings cut at both sides through which a photograph may be taken.

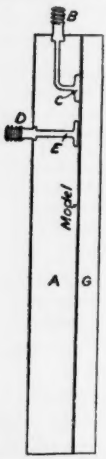

FIG. 1.—The Air-Tight Containers of the Hele-Shaw Apparatus.

Since the density of the fluid is constant, it follows that the same volume must pass across each successive section of the model, and since the number of stream-lines remains unchanged, the relative velocity of the flow at any point is inversely proportional to the distance between successive lines. 'Knowing this relative velocity and the distance travelled along each line, it is a matter of simple calculation to determine where a given element of the initial section will have arrived after successive intervals of time. This method of analysis has been used in connection with the work on the flow in the wire-drawing die which is described in Part II., Section (*c*).

Part II.

An attempt has been made in the following pages to apply this apparatus to the examination of some simple forms of drop stamping, to the drawing of tubes and mainly to an investigation of the nature of the flow during the drawing of wire. So far as the present consideration is concerned, no differentiation is to be made between hot- and cold working, since stream-line flow occurs in both instances, different though these processes are in their effects upon the properties of the deformed metal.

Section (a).

In Figs. 5 and 6 (Plate XII.) are shown the macrostructures of two simple forms of drop stamping. In Figs. 7 and 8 are reproduced photographs of the flow through similar contours in this apparatus, and the essential similarity of the two sets of illustrations will be at once appreciated. Too exact concordance must not, however, be expected, owing to flaws which may occur in the material, and which might interfere with the flow; to variations of the surface friction, the importance of which has been shown by Siebel and Osenberg[2]; and to the discontinuous flow due to the individual blows of the hammer.* From a comparison of these two sets of illustrations it would appear that the Hele-Shaw apparatus is capable of yielding, at any rate qualitatively, very clear information regarding the manner in which metal may be expected to flow when stamped between dies.

Section (b).

The recent work by Pomp and Schylla[3] on the flow of metal during tube drawing may be employed to enquire to what extent the flow there obtained corresponds to that which occurs in the apparatus now being described. A typical photograph from the German work and the corresponding Hele-Shaw flow lines are reproduced in Figs. 9 and 10 (Plate XIII.), and again the close analogy between the two will be apparent. In Fig. 11 the two figures have been superposed and the close similarity will be at once seen.

The authors have used this apparatus in the examination of the flow during the extrusion of metals and have obtained stream-lines which are strictly comparable with those which have been obtained by other investigators. Since, however, this work is of more importance in non-ferrous than in ferrous metallurgy, the results are not reproduced here, the fact being mentioned merely as some further indication of the way in which this type of investigation can be applied to the examination of an industrial working process.

* The apparatus is quite capable, however, of investigating the effect of *known* flaws, as Figs. 3, 21 and 22 will show.

Section (c).

The authors' interest in this method of examining flow was aroused in the first instance as a result of their work on the drawing of wire, and the major portion, and in fact the whole of the quantitative section of the investigation, applies to this particular form of deformation.

Since, however, in the Hele-Shaw apparatus the fluid is forced through the die, whilst in the drawing of wire the metal is pulled through, a certain doubt might arise as to the justification of comparing the results in the two cases. So long, however, as the flow is of a pure stream-line type, the direction in which the pressure varies should not hydro-dynamically have any effect upon the nature of the flow lines. To satisfy themselves that this was in fact the case, some preliminary experiments were carried out in which the glycerine was drawn through the die by suction from below instead of, as is normally the case, by forcing it down from above. The corresponding results are shown in Figs. 12 and 13 (Plate XIV.), and it will be seen that the flow has, as far as can be seen, been identical in both instances. To confirm this conclusion, the two photographs were analysed quantitatively in the way outlined above. In Fig. 2 is shown what may be called the " wave-fronts," and those (in full lines) due to the suction and those (dotted) obtained in the normal manner, fit into each other perfectly. It may be concluded, therefore, that the picture of the stream-line flow in wire-drawing obtained with the conditions normally used, is not in any way affected by the direction in which the pressure varies.

Since the present paper is mainly concerned with a general description of the use of this apparatus in investigating the flow of metals, it will be sufficient to describe a single instance in which the stream-line picture has been fairly completely analysed. In Fig. 14 the stream-lines for a radial taper die and a reduction of diameter of 50 per cent. are shown, together with the lines of equal velocity of flow on the one hand and what may be termed successive wave-fronts on the other. The lines of equal velocity can be obtained directly from the measurement of the relative distance between successive lines. Knowing this and the distance which a particle must travel along a given stream-line it is, as has already been pointed out, a matter of arithmetic alone to determine where a given particle will have arrived after the expiration of a given time interval. The dotted lines in Fig. 14 represent these loci. Among the features of interest it will be noted in the first place that the effect of the die is felt some considerable time before the metal actually enters it, particularly towards the centre. This is in complete agreement with practically all the work which has been done on wire-drawing—for instance the photo-elastic investigation of Thompson and

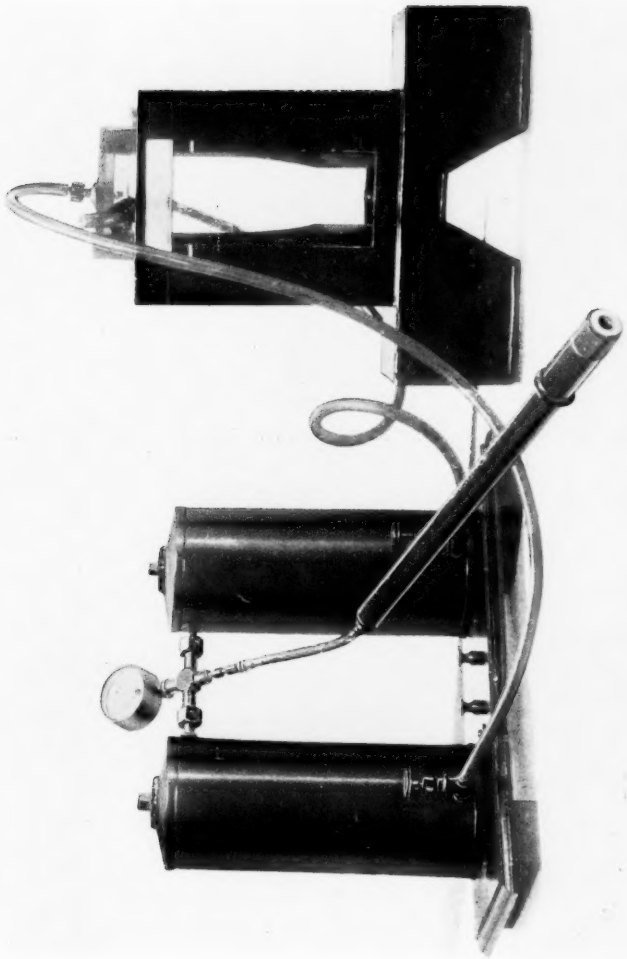

FIG. 4.—The Hele-Shaw Apparatus.

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To face p. 120.

FIG. 5.—Simple Drop-Stamping Macrostructure.
(See Fig. 7.)

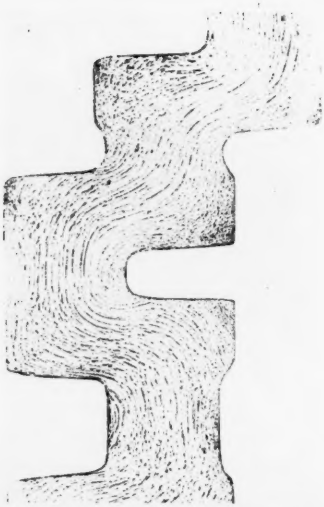

FIG. 6.—Simple Drop-Stamping Macrostructure.
(See Fig. 8.)

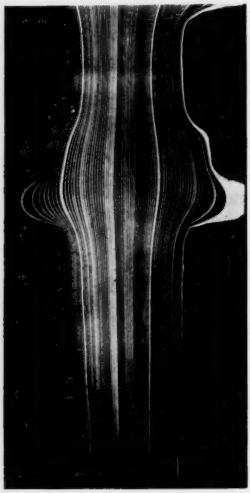

FIG. 7.—Flow through Contour of same shape as Fig. 5.

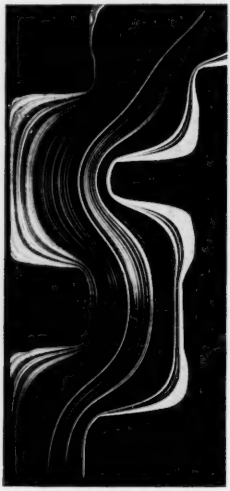

FIG. 8.—Flow through Contour of same shape as Fig. 6.

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FIG. 11.—Negatives of Figs. 9 and 10 superposed, to show the close similarity.

FIG. 10.—Hele-Shaw Flow Lines corresponding to Fig. 9.

FIG. 9.—Flow of Metal in Tube-Drawing (Pomp and Schylla [23]).

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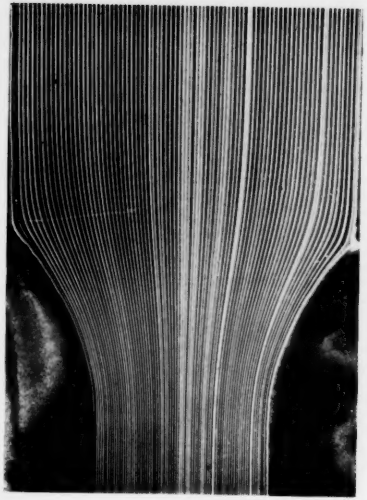

FIG. 12.—Effect of *Drawing* the Glycerine through the Die. (See Fig. 13.)

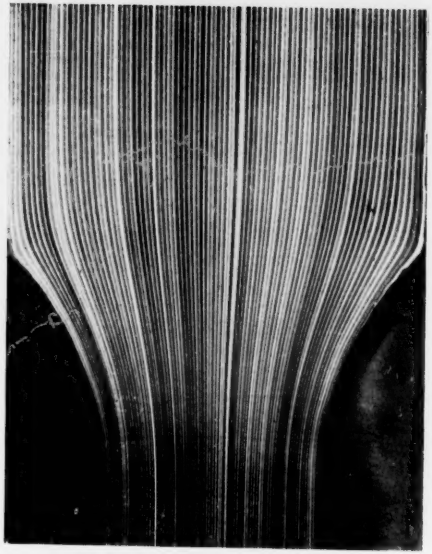

FIG. 13.—Effect of *Forcing* the Glycerine through the Die. (See Fig. 12.)

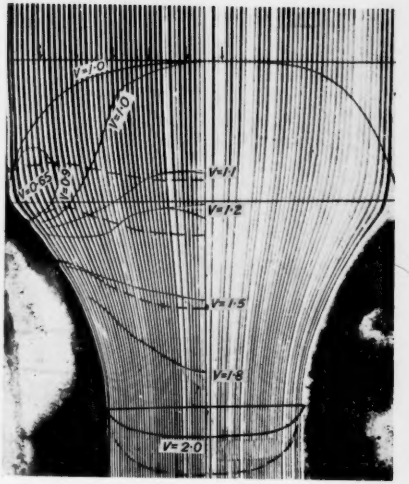

FIG. 14.—Stream-Line for a Radial Taper Die.

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PLATE XIV.

FIG. 15.—One of the Specimens used by Taylor and Quinney.

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FIG. 16.—Wave-Front.

FIG. 17.—Wave-Front.

FIG. 18.—Wave-Front.

FIG. 19.—Superposition of Hele-Shaw Stream-Line on Macro-etching of partly-drawn steel bar of same contour.

FIG. 20.—Flow in a Bell-Shaped Die.

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steel bar of same contour.

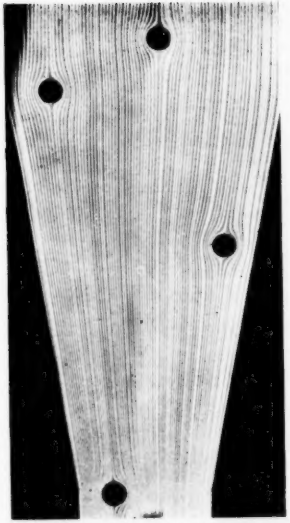

FIG. 21.—Effect of "Inclusions."

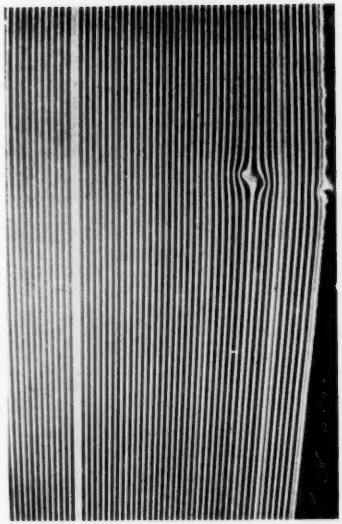

FIG. 22.—Flow round a Particle of Dirt.

FIG. 23.—Effect of Asymmetric Flow, representing "cast" in a wire.

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Barton[4] and the work of Taylor and Quinney.[5] Towards the surface the increased width between the lines indicates a definite dim nution in the velocity of the movement as the fluid approaches the die, which in the particular contour now under consideration falls to less than 0 65 of the initial velocity of the metal at the surface at the point of entrance. This retardation is again in agreement with the conclusions reached in the photo-elastic investigation already mentioned. From this point the velocity, which so far has been decreasing, commences to rise again,

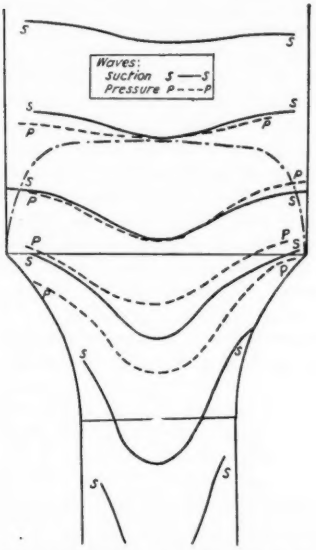

FIG. 2.—Analysis of Photographs in Figs. 12 and 13.

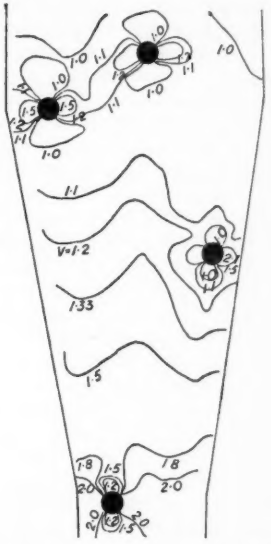

FIG. 3.—Effect of "Inclusions."

reaching its maximum on the surface at the point of exit from the die. On the centre line there is no indication of this retardation, the velocity increasing very slowly at first and then more and more quickly as the die is traversed. Passing on to consider the wave-fronts it may be pointed out in the first place that these are characteristically bowed, the depth of the curvature increasing with the reduction, as Fig. 14 shows. In Fig. 15 (Plate XV.), which represents one of the specimens investigated by Taylor and Quinney, this characteristic bow shape has also been found, and the close agreement between the Hele-Shaw results and those obtained in actual wire-drawing is, in the opinion of the authors,

extremely strong confirmation of the accuracy with which the former corresponds to practical conditions.

A considerable amount of time was spent in endeavouring to obtain direct photographs of these wave-fronts. The difficulties of getting initial uniformity of flow were, however, extremely great and this aspect of the investigation, beyond providing a qualitative confirmation of the conclusions to be drawn from Fig. 14, was given up. In Figs. 16, 17 and 18 (Plate XVI.), however, are reproduced some of the best of these photographs, from which it will be seen that the general shape of the flow lines deduced mathematically from the stream-line measurements does correspond extremely closely with that actually obtained. The increased depth of the bow as the reduction proceeds is very clearly shown.

In an attempt to discover to what extent the Hele-Shaw stream-lines did in fact correspond with the flow of the metal, a fairly thick bar of steel, after being partly drawn down was cut off and knocked back. From this a section was produced and macro-etched. The photograph of the corresponding stream-line was prepared from an identical contour and of exactly the same size, and from the two negatives the composite photograph in Fig. 19 was produced. So far as the most careful examination that the authors have been able to make goes, it would appear that the flow is identical in each case. A similar result has been obtained for other wires which show badly segregated cores. Since, however, the flow of the segregated and of the purer portions may not, and in fact probably do not, coincide, these pictures have not been reproduced here.

The work of Thompson and Barton suggested that a die with a contour the reverse of the "radial taper," which in that paper was referred to as a " bell-shaped " die, should permit of flow such that an initial plane section prior to drawing remains almost a plane afterwards. In Fig. 20 is shown the analysis of the result obtained from a die of this character, from which it will be seen that the amount of curvature of the dotted wave-fronts is exceedingly small. A more appropriately shaped die would lead to complete flattening out. The authors are not concerned here with the practical question of the advantages and disadvantages of such a die ; they are, however, concerned with pointing out that the conclusions of this work do tally exceedingly well with those of the other and very different investigation.

The importance of inclusions in the metal in the drawing of wire needs no emphasis, and it appeared to the authors interesting to attempt to examine the effect of such inclusions on the flow which takes place. In Fig. 21 (Plate XVII.) and Fig. 3 are reproduced the stream-lines and the lines of equal velocity of flow resulting from attaching circular discs of paper to the glass wall of the apparatus. It is clear that this does not correspond exactly

with the actual case, where the inclusions flow with the metal itself, and it is possible that some degree of exaggeration of the irregularities introduced may have occurred. The work is, however, not without its interest in emphasising the distortion of the stream-lines and the extreme irregularity of the velocity of the flow, not only in the immediate neighbourhood of the inclusions themselves but for some very considerable distance away. Qualitatively, the pictures are also of interest in another direction. Changes of direction of the stream-lines and changes in the relative distance apart involve lateral stresses. Fig. 3 in particular will show that even where the inclusion is some distance below the surface of the metal, the nature of the stream-lines indicates appreciably increased pressure upon the dies. So far as the authors are aware this is the first time it has been suggested that the excessive wear of a die due to the inclusions in the metal being drawn may not be due solely to the scratching of the die surface by the hard inclusions themselves, but may also be due to increased pressure resulting from inclusions which never actually come in contact with the die at all.

Where the inclusion is flowing along with the metal itself there is still strong evidence of the distortion of the stream-lines. At times small air bubbles were accidentally introduced along with the glycerine, flowed down through the die along with the latter and produced very great distortion in their neighbourhoods. Fig. 22 shows an enlarged picture of the stream-lines around a small particle of dirt which on one occasion was photographed during its passage through the die. Although this particle was flowing with the glycerine itself the distortion of the stream-lines in its neighbourhood will be clearly seen.

In view of the importance, from the point of view of the "cast" of the wire, of the angle at which the metal emerges from the die, a typical example of the flow which takes place when the drawing is asymmetrical is reproduced in Fig. 23. The differences in the deformation on the two sides of the wire are very clearly shown.

Conclusions.

The authors have in the present paper shown how the Hele-Shaw apparatus may be employed to obtain not only a qualitative picture of the manner in which metals flow during deformation, but also the extent to which quantitative data can be derived. The latter part of the work, which is concerned entirely with the drawing of wire, results in the prediction of wave-fronts which are in complete agreement with those obtained by actual drawing. Further, the results indicate that the effect of the contour of the die on the extent to which the centre flows in advance of the surfaces coincides with that obtained in a totally different manner by Thompson and Barton, and with the general

experience of those concerned with the production of wire itself. Without making too strong claims at the present time, it is felt that the cumulative evidence presented is sufficient to justify the belief that, qualitatively, and at any rate to some considerable extent quantitatively, the method which has been described agrees with actual fact. The ease with which the stream-lines can be obtained makes the method qualitatively of real practical utility, although it should be pointed out that the analysis of the results where quantitative information is required, is a distinctly tedious process. The fact that a two-dimensional solution is all that can be obtained by this process must be borne in mind, but in those cases where the deformation is symmetrical this is probably a faithful reproduction of the flow, at any rate on sections parallel to the axis of the section.

It should, perhaps, be pointed out that such experiments with a fluid cannot reflect any aspect of deformation which is due to the vectorial, crystalline properties of the metal. For example, Taylor and Quinney have shown, and this has been confirmed by Siebel,[6] that when an unworked rod is drawn through a die, the shear occurs almost entirely in the first pass, and thereafter remains essentially unchanged. As a result, the shape of the wave-front after the first pass does not coincide with that due to subsequent passes of the already cold-drawn material. It is clear that the behaviour of the latter cannot be exactly duplicated by the flow of a fluid; all that the latter can do is to provide a picture of the deformation of the original, isotropic metal.

The authors would like to express their deep appreciation of the great help they have received throughout from Mr. W. Ashelby. Acknowledgments are also due to the Editor of *Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung*, Düsseldorf, for permission to reproduce Fig. 9 (taken from the work of Pomp and Schylla[3]).

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DISCUSSION.

Dr. CECIL H. DESCH, F.R.S. (Vice-President, Teddington, Middlesex), in opening the discussion, said he was sure all the members were interested in the application of the very beautiful Hele-Shaw method to such a problem as the authors had investigated, and the experiments were certainly striking. In the references to the previous literature he saw no mention of the work of Gulliver, which rather surprised him. Gulliver, some years before the war, published a paper[1] in which he applied the Hele-Shaw method to the examination of the flow in test-pieces. He made his dies to represent the shape of test-pieces and then applied quantitative calculations, as the present authors had done, to determine what should be the plane of fracture, and found a very good correspondence with the actual results. Having done that, he investigated the effect of sharp and rounded notches of various angles, and the Hele-Shaw lines again corresponded very closely with the actual deformation and fracture found in test-pieces. The paper in question did not seem to be well known, perhaps because people did not often see the *Proceedings* of the Royal Society of Edinburgh, but it was a valuable piece of work.

The question arose, how close was the analogy between the flow of a viscous liquid and that of a plastic solid during deformation? The authors had shown that the analogy was close over a certain range, but there must, of course, be differences because of the enormous difference in viscosity between even the most viscous liquid and a solid. For example, reference was made to the fact that real turbulent flow was never obtained in a plastic solid, whereas turbulence could easily be obtained in a liquid, even a viscous one. Over the range examined in the paper, however, the analogy seemed to hold.

The authors attempted to test whether push and pull were really equivalent. They could not be so over the whole range, as was shown by extrusion. In discussions on extrusion before the Institute of Metals the fact became conspicuous that if one streamlined the die in an extrusion press the metal would not pass through at all; it jammed. The die must have a perfectly sharp right angle, and then the metal would flow, as in Tresca's experiments. Some of Tresca's early experiments on lead and heated iron were made with an extrusion apparatus using sharp right-angled dies. The flow was then not stream-line flow but differed widely from it, and the question arose, at what stage of deformation did the difference between stream-line flow through a wire-drawing die and non-stream-line flow through an extrusion die come in?

[1] *Proceedings of the Royal Society of Edinburgh*, 1909-1910, vol. 30, p. 38.

There must be some critical point at which the analogy broke down; so that the present paper, which was extraordinarily interesting, opened up some new problems which seemed to him very well worth while investigating.

He raised those points because the authors would probably think it worth while, since they had obtained the necessary manipulative skill to use the really beautiful method adopted, to go on with the experiments.

Mr. T. HENRY TURNER (Doncaster) pointed out that however true the authors' pictures might be for the flow lines seen in forgings, there could be no doubt that cold mild steel commenced to deform in quite a different manner.

Lüders lines and Fry etching showed that at the yield point wedges of deformation grew at approximately 45° to the lines shown by the authors. There must therefore be some change in mild steel after the initial deformation seen in Lüders lines if it later flowed in stream-line form. To examine stress concentrations below the yield, one should paint the surface with suitable lacquers and watch the failure of the lacquer where the metal stretched. To examine actual deformation just above the yield point, Fry's etching was employed. For finding the flow after severe deformations, deep etching of a section was most reliable for big forgings.

The authors' experiments suggested that for wires, tubes, small drop forgings and other such articles where deep-etched cross-sections might be less informative, the Hele-Shaw apparatus might be of assistance to investigators.

Dr. E. W. FELL (Manchester) said he would confine his remarks to the summary of the paper. In it mention was made of flow-lines actually found in the metals themselves; and, regarding the Hele-Shaw apparatus, it was stated "it would, therefore, appear that it is legitimate to employ this type of apparatus where the mode of flow during any process of deformation is required to be known." He would suggest, however, that the lines found in the macrostructures of any general type of forging, or of cold-worked steel, were not really lines of flow. He thought they might be a sort of qualitative indication of the flow which had taken place; that they were not lines of flow might be seen by an illustration. Consider a bar to be forged and in it two tiny particles, say two non-metallic inclusions, a very small distance apart. The forging operation (the actual deformation) might be divided into small or infinitesimal stages, in each of which the line joining the two particles would suffer displacement consisting of a translation parallel to itself and a rotation relative to its former position. This occurred in each infinitesimal stage, hence the position of the

line at the end of forging, as, for example, in Figs. 5 and 6, represented a rather complex displacement. Thus the lines in the two figures were not lines of flow, that was, the tangent to the lines at any point was not the direction of motion of the metal at that point. The lines for the steady motion of glycerine, shown in Figs. 7 and 8, were lines of flow.

The matter, he thought, might be visualised in another way. There seemed no reason for supposing that the flow in the channels of glycerine should coincide with the macrostructures of forgings happening to have the same sectional shape as the channels; the two states arose from quite different causes. The lines of flow in the glycerine would not accordingly be expected to coincide with the lines in the macrostructures.

The case of a partly-drawn tube or wire was a particular one, and the inscribed lines in the metal, apparently shown in Figs. 9 and 15, and the lines in Fig. 19, suggested lines of flow, assuming that motion was steady through the die and that no further distortion occurred subsequently. It seemed, in studying the flow, that consideration should be given to the fact that the boundary of the channel in the Hele-Shaw apparatus was quite different from the boundary (surfaces of contact of the metal and die) in the drawing operation.

Drop-forging was a case of general forging, though it might take place more quickly and the disposition of the metal might be better controlled. The operation, however, could be divided into small stages, as described above, and similar remarks would apply.

He expressed the hope that his remarks might be of interest to the authors of the paper.

Mr. J. D. Hannah (Manchester) said that in reading through the authors' exceedingly interesting paper, in which what appeared to be a valuable analogy had been drawn between the stream-line flow of liquids and what was referred to as the flow in forged parts, he found in section (*c*) of Part II. a paragraph in which it was claimed that by varying the arrangements for applying pressure to the liquid in the Hele-Shaw apparatus an analogy had been obtained with two cases which occurred in handling metal, the first where the metal was forced through a die and the second where it was drawn through it, as in the case of wire. The authors stated that by applying suction below the die they had obtained an analogy with the position that obtained where wire was drawn through a die. Personally, he felt unable to follow the reasoning there. In applying suction, presumably the authors had simply removed atmospheric pressure from beneath the liquid, leaving it still applied at the top. He was unable to see that that differed fundamentally from applying a pressure greater than atmospheric above the liquid; it seemed to him a purely relative

matter and not fundamentally different. In drawing wire one could understand that the conditions were, in fact, different, because the wire was under tension ; but it did not seem probable that the liquid could, in fact, sustain tension.

Dr. W. H. HATFIELD, F.R.S. (Vice-President, Sheffield), congratulated the authors on their very interesting paper. Professor Thompson, he said, came from Sheffield, and the work he was now doing in Manchester was worthy of his old city.

Personally, he thought the last speakers had been inclined to credit the authors with reading too much into the experiments described. Taking Figs. 6 and 8, he thought the analogy between the flow through a contour and the drop-forging was very complete, and this lent itself very well indeed to the determination of a lot of comparative and empirical data which would be very helpful and instructive, always bearing in mind that one had two dimensions and not three. He would like, therefore, to congratulate the authors on a very useful contribution.

On the other hand, when one examined Fig. 3 and saw the appalling effect of those inclusions, he could not help thinking that some of his friends must have been supplying some of those samples. Speaking seriously, if one were working in three dimensions instead of two, and with a material of the viscosity of the heated steel, he ventured to say that the results would be in no sense comparable.

A point which occurred to him in studying the paper was that, for the purpose of the present discussion, the authors visualised those inclusions as fixed, immovable bodies of certain intrinsic characteristics which remained constant. The point they raised was a very important one, especially in regard to inclusions in drop-forgings. On the other hand, when one considered the inclusions as "glasses," and when the steel was heated and became plastic, it was a fair inference that the characteristics of the slag inclusions were modified at the same time. At any rate, the slag inclusions were so considerable in number but so small in dimensions that he questioned very much whether that particular analogy was not far too much overdrawn, and in a sense somewhat mischievous ; it might create an impression in the mind of users of steel, for instance, that inclusions in steel were really a disability.

That led him to make this observation. The authors dealt with cold steel as well as with hot steel ; in other words, they dealt with steel which presumably had a ferritic matrix of a highly plastic nature, but disseminated through that matrix was the carbon which gave strength to that material in the form of hard, brittle carbide plates or nodules. It might be suggested to the authors that in the cold it would be very difficult to discriminate

between the behaviour of the hard carbide particles to which the ultimate properties of the steel were due and the hard metallic inclusions. That touched on a point which was of very great interest. If the strength of steel as affected by the carbon content was really due to the fact that the hard particles affected the manner of slip, it was quite possible that those inclusions which the authors treated so severely might have a similar desirable effect. He did not wish to excuse some of the products that, he believed, were put on the market in the ferrous industry, but at the same time there was something that might be considered seriously in what he had just said.

In conclusion, he would like to congratulate Mr. Herbert and Dr. Thompson on a very constructive piece of work.

Dr. A. McCance (Member of Council, Glasgow) said that, considering Fig. 3 a little further, he thought there was a danger in applying too closely the method of accurate measurement to the spacing of the lines. If reference were made, for example, to the lower particle, it would be noticed that the wings in the case of that particle were lengthways, longitudinal, whereas in the case of the other three they were at right angles to the direction of flow. He thought there must be an error of measurement owing to the spacing of the lines being too coarse, because in the photograph it would be seen that the dead areas for each hole were in each case lying along the axis in the direction of flow, where they ought to be; *i.e.*, maximum stress at the hole was always at right angles to the pull, whereas according to the measurements made in the case of the lowest hole it appeared that the stress was actually at right angles to the direction in which it ought to be.

With regard to the discussion which had been taking place as to whether there was not a real analogy between flow-lines and the effects discovered in etching and drop-forging, he thought the point had been overlooked that while the material was flowing it had reached a steady state, and so aspects of flow did not enter into the matter. What was shown was a picture where the tangent to the direction of motion at any particular point had remained constant, and consequently the picture shown really indicated displacement, and to that extent it was exactly analogous to any movement that had taken place or any lateral displacement in a bar with longitudinal lines in it.

The only case where it would not be applicable was that to which other speakers had referred, when with plastic solids the displacement took place in slip movements along shear lines at 45°. If those movements were too coarse one could imagine a zigzag effect, and that could not be analogous to a steady flow line. But where those steps were very small, with a multitude

of very small displacements, then he thought the analogy was quite justified, and the authors' method was applicable.

CORRESPONDENCE.

Mr. E. A. ATKINS (Warrington) wrote that whilst the authors very ingeniously showed that certain *results* appeared to be parallel, it was not easy to grasp that the movement of the metal in the drop-stampings was at all similar to the flow of liquids in the Hele-Shaw apparatus. In the one case the material simply closed up and showed the disposition of the heterogeneous fibres, and in the other it followed the distortion of liquid stream-lines in a tortuous conduit. However, the results might be of practical use in showing up the planes of weakness in stampings.

It was not at all surprising that the results of forcing the glycerine through a die were the same as when it was passed through by suction, as the latter case would be the same as the former, the atmosphere supplying the pressure. The effect when an effort was made to push a rod through a die was quite different from when it was pulled through. In the first case the metal jammed itself in the die, whereas in the second the metal at the middle was pulled away in advance, thus allowing the outside to collapse under the pressure from the die walls. In extruding metals the dies were so shaped as to avoid jamming by allowing the core of the rod to move forward in advance of the skin, thus bringing about a similar condition to the drawing of wire.

Mr. A. B. WINTERBOTTOM (Trondheim, Norway) wrote that Hele-Shaw's stream lines *in thin layers of a viscous fluid* were found to be identical in form with the lines *for two-dimensional flow of a perfect non-viscous fluid*, which were calculable for symmetrical cases.[1] Stokes[2] showed that that result was to be expected. Flow models should therefore be applicable to the study of deformation insofar as it was similar to the flow of a perfect fluid not complicated by shear and friction effects, and provided that two-dimensional models satisfied symmetry requirements. Discrepancies due to this latter factor might only be serious when the initial and final sections were not geometrically similar and the reduction was large. Thus the diagram (Fig. A) showed calculated final flow distributions for two hypothetical cases of extrusion, where the sections were geometrically similar in the one case and not so in the other. In both cases two-dimensional flow models would give uniform distributions in the exit channel, a

[1] "Hydrodynamics," Lamb, p. 61.
[2] *British Association Reports*, 1898, pp. 143–144.

result which was in error for the second case. The flow lines must therefore be distorted throughout their courses in that case.

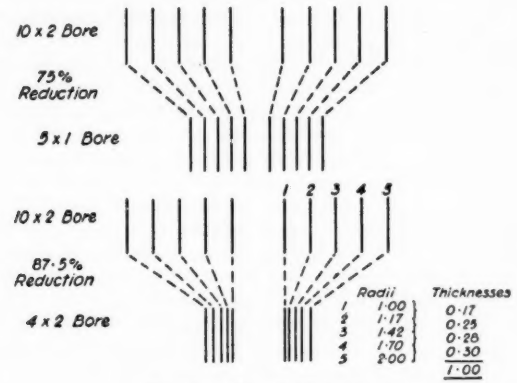


FIG. A.—Flow in Extrusion.

It would be interesting to know whether the surface slowing-up and bulging of the fibres (longitudinal compression and transverse expansion) observed at the die entry in the model experiments was confirmed in like degree in actual drawing.

AUTHORS' REPLY.

Professor THOMPSON, in reply, said the authors were extremely gratified by the discussion that their paper had aroused. They were much indebted to Dr. Desch for his reference to the paper by Gulliver, of which they had been quite unaware. This they had now added to their list of references to make it complete.

Both Dr. Desch and Dr. Hatfield raised the question of the influence of the viscosity. The answer was clear; provided that the flow was stream-line in nature, the viscosity was without influence. The authors did, in the early days, one or two experiments on extrusion and thought the results explained why it was that in satisfactory extrusion it was necessary to have a sharp angle to the die, and not, as one might have expected at first sight, a stream-line die. At the sharp angle the stream-lines were pressed very close together, an indication that the pressure on the metal at that point was extremely high. As a result, flow of the material took place relatively easily. On the other hand, in the case of a stream-line die the stream-lines were relatively

far apart, indicating a relatively low lateral pressure; and he thought, therefore, on more mature consideration it was not surprising that the sharp angle was required.

With regard to Mr. Turner's contribution, the authors, he agreed, had not so far at any rate been able to duplicate the cone or pyramid that one got at the commencement of flow in a metal. On the other hand one could, by using the quasiquantitative method that the authors had applied in the case of wire, get something which was rather close to it, and he was not at all sure that further work would not make it possible to get at any rate some sort of rough approximation.

With regard to Dr. Fell's contribution, he thought Dr. Fell was looking at the matter more fundamentally than the authors had ventured to do. If one went down to atomic displacement, one got a state of affairs which it was not possible, of course, even approximately to duplicate in this apparatus. On the other hand, if one considered the position which inclusions had taken up after a certain deformation of the metal, *i.e.*, looked at the problem in a broad way, it seemed rather difficult not to believe that the location of the inclusions—and therefore presumably the metal round those inclusions—represented the position into which the metal had flowed. He was inclined to think that Dr. Fell was looking at the matter much more fundamentally than the method justified, but he would like to emphasise what Mr. Herbert very properly said when introducing the paper, that the authors made no dogmatic claims in any direction.

Dr. Hatfield's interesting and, if he might say so, characteristic contribution raised important points. With regard to the effect of inclusions, he wished to point out that they had considered these only from one point of view, namely, that of their effect on wire rod. He believed that all drawers of wire would agree that inclusions were bad. With regard to the nature of the inclusions, it was immaterial whether these were in part glassy, as Dr. Hatfield very properly pointed out, or crystalline as in the case of one of the most objectionable of all, alumina.

There was only one other point with which he would deal, namely, with regard to the tube. In no case had the authors been able to get a more striking confirmation of the agreement of the flow obtained in the apparatus used and the flow that actually occurred than in the case of tube drawing. They took the figure obtained by scribing parallel lines on the undrawn tube which had been obtained by Pomp and Schylla, and made a die which as nearly as possible agreed with this contour. Prints of each were made on semi-transparent paper, and by superposing the two—the authors' own Hele-Shaw stream-lines and the scribed lines obtained by Pomp and Schylla—they made a composite print which he was able to produce. In every particular

the two sets of lines were parallel from beginning to end. That was, he thought, the best example of the correspondence between the Hele-Shaw lines and the actual flow which up to the present they had been able to obtain.

Mr. HERBERT, who also replied, associated himself with Dr. Thompson in thanking those members who had obviously taken some interest in and studied the paper. He agreed with Dr. Desch that the important point was the range through which the analogy held. Dr. Fell seemed to mean that the lines were not really flow-lines when they were a series of small movements, but if one integrated all those small movements one practically got a flow-line, or at any rate a line which was comparable to that flow-line. Nobody had referred to the authors' suggestion on the penultimate page of the paper as to the effect of inclusions in the wire on the wear of the die. Possibly no one liked to commit himself to a definite opinion, and the authors entirely shared that view.

In further reply to the discussion and correspondence, the AUTHORS wrote that it should be pointed out, in connection with both the remarks of Dr. Fell as well as those of others, that the work described in their paper did not attempt to deal with the mechanism of the deformation of metals, being merely concerned with an investigation of the final positions taken up after flow had taken place. It was clear that atomic movement within a crystal could not possibly be expected to be analogous to the viscous flow of glycerine. What the authors had attempted to show was that after deformation had occurred it was possible, from the work on the stream-line flow of the fluid, to predict the situation of any particular particle of the metal.

Mr. Hannah and others had pointed out that in the work on the suction of the glycerine through the die, the direction of the pressure variation was identical with that when the apparatus was employed in the normal manner. The authors were not, however, particularly concerned with this matter, the tests recorded being carried out merely to do what was possible to meet any possible criticism that the drawing of the wire might differ essentially from the state of affairs when the glycerine was forced through the die. Although under certain circumstances it is well known that liquids could sustain tensile stresses of no inconsiderable magnitude, it was not suggested that in the present work the liquid was drawn through the die in the same sense as would be a wire.

With regard to Mr. Turner's contribution, Gulliver[1] had shown how exceedingly closely Lüders lines could be accounted

[1] *Proceedings of the Royal Society of Edinburgh*, 1909-1910, vol. 30, p. 38.

for from Hele-Shaw stream-line pictures. It would appear, therefore, that even under stresses which only just exceeded the elastic limit the actual deformation and the viscous flow in the Hele-Shaw apparatus followed closely similar paths.

The authors wished to thank Dr. Desch for his suggestion that the analogy between the flow of metals and that in the Hele-Shaw apparatus might hold only over a certain range, a suggestion which, if possible, they would investigate.

They were glad to have Mr. Winterbottom's opinion, which agreed with the authors' views as well as with those of Dr. McCance, that inherently the Hele-Shaw method was a suitable means of investigating the two-dimensional flow of a metal. They were not clear as to the manner in which Mr. Winterbottom had calculated the results from which his diagrams were drawn, and could not, therefore, express any opinion upon them. With regard to the question as to the bulging of wires immediately prior to their entrance into the die, this in certain cases could be shown to be an experimental fact. There were, however, others in which no such increase of diameter appeared to occur.

That the flow at the surface of the die could not be strictly analogous to that at the surface of a metal during its passage through a wire drawing die was, of course, perfectly clear. Any discrepancies due to this cause, however, probably extended to such a small distance from the surface that for most purposes they could be entirely neglected.

EXAMINATION OF THE SURFACE OF TINPLATE BY AN OPTICAL METHOD.*

By W. E. HOARE, B.Sc. (The International Tin Research and Development Council), and BRUCE CHALMERS, Ph.D. (The Sir John Cass Technical Institute).

Summary.

The surface of tinplate has been examined by an optical method involving the formation of interference fringes between an optical flat and the tin surface.

The method enables a distinction to be made between hollows and reliefs in the surface, and the depth of an irregularity is easily determined by counting the fringes under visual observation.

Sections of the more significant surface irregularities have been plotted on an enlarged scale.

The authors point out the possibilities of the method in other fields of investigation.

The usual hot-dipped tin-coating on steel is extremely thin, and methods of examination involving sectioning the edge at right angles are not sufficiently convenient to be generally applicable.

A tin-coating equivalent to $1\frac{3}{4}$ lb. per basis box,† which is fairly common and satisfactory practice, is equal in thickness to approximately 0·0001 in. or 0·0025 mm. of tin on each side of the steel basis sheet. It will be seen that at a magnification of 1,000 diameters the tin-coating will, in this case, appear as a layer only $\frac{1}{10}$ in. in thickness, in which dimension it is hardly possible to observe details of structure with any clarity. One of the authors, in a previous publication before this Institute,[1] has outlined some of the difficulties encountered and described methods by which they may in some measure be overcome.

The methods of examination described formerly, however, were chiefly designed as means of observing the structure of the $FeSn_2$ layer between the steel base and the coating metal, and suffer from the disadvantage that it is impossible, by their use, to determine with any accuracy the contours of the surface. It is with this consideration in view that the method described here has been devised. The technique involves the formation of

* Communication from the Sir John Cass Technical Institute, London, received June 3, 1935.

† 1 basis box = 31,360 sq. in. of tinplate.

interference fringes between an optically flat surface and the surface of the tinplate, under vertical illumination with monochromatic light.

Theory.

The assumption is made that the specimen to be examined is vertically illuminated through an optical flat with monochromatic light of wavelength λ at the site of an irregularity of depth d. (See Fig. 1.)

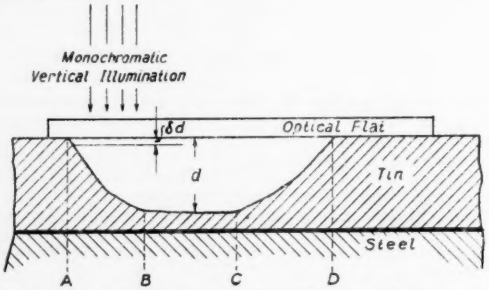


FIG. 1.—Diagrammatic Representation of the Method of Formation of Fringes.

Then that part of the surface of the specimen to the left of A will appear uniformly bright or dark, that between A and B will show a series of interference fringes, and that between B and C will again be uniformly bright or dark. The same conditions obtain at the other side of the hollow.

Along AB:
The 1st bright band will occur at $d = \lambda/4$
1st dark " " " " $d = \lambda/2$
2nd bright " " " " $d = 3\lambda/4$
and so on.

If N is equal to the number of complete fringes between A and B,

$$2N = 4d/\lambda, \quad N = 2d/\lambda.$$

When $\lambda = 546\mu\mu$ (*q.v.*)
$N = (d/273) \times 10^7$ . . . (i)
and $d = 273N \times 10^{-7}$ . . . (ii)

The number N in the case of ordinary commercial grades of tinplate is found to be of convenient size.

Thus in a 2 lb. grade tinplate, having approximately 0·00012 in. of coating tin on each side of the steel basis sheet, the maximum value of d will obtain in the case of a true pore,* *i.e.*, when

$d = 0{\cdot}00012$ in.
from equation (i)
$N = 0{\cdot}00012 \times 2{\cdot}54 \times 10^7/273.$
$N = 11.$

* A pore is defined as a local microscopic exposure of the steel base.

The character of an irregularity (convexity or concavity) can be simply determined by applying a slight increase of pressure to the optical flat. As the clearance between the two surfaces decreases, the line of contact moves downwards across the contours. In the case of a hollow, therefore, the fringes will move inwards and disappear at the centre, the reverse obtaining in the case of a convexity.

Apparatus.

A sketch of the very simple apparatus required is given in Fig. 2.

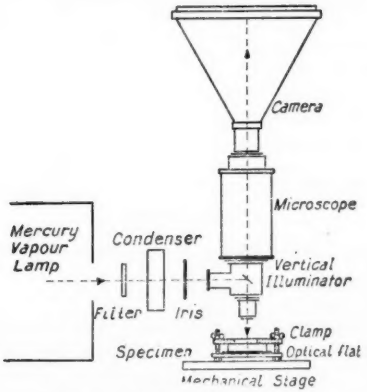
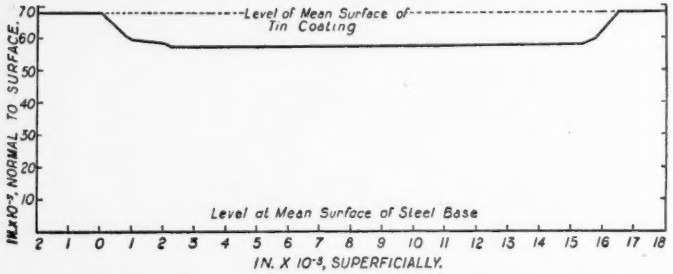


Fig. 2.—Interference Apparatus.

Monochromatic light is conveniently obtained from a mercury-vapour lamp equipped with a Wratten No. 62 filter, which excludes all wavelengths but the mercury line at 546$\mu\mu$. This line is in the green part of the visible spectrum and therefore quite suitable for photographic record. The light beam is taken through a condensing system and iris diaphragm to an ordinary vertically illuminating metallurgical microscope, equipped, if necessary, with a camera attachment. The optical flat, which should be as thin as possible (*e.g.*, 0·1 in.) in order to provide for a short working distance, and therefore a high magnification, is placed over that part of the surface of the specimen to be examined and held in position with thumbscrews to regulate the pressure. It is convenient also to use a mechanical stage, as this facilitates moving the specimen, and, together with a cross-wire, is essential when the shape of an irregularity is being accurately plotted.

In the initial stages of this work the authors found it convenient to study the defects known as grease marks. These marks are

clearly visible without etching, and in some cases are of fairly large size, giving bright and well separated fringes at comparatively low magnifications. The interference effect caused by such an irregularity is shown in the photomicrograph (Fig. 8, Plate XVIII.). Using the pressure method, it was established that this mark, and all other grease marks inspected, are hollows in the tin coating, although not necessarily extending through to the compound layer or the steel base. A section through the grease mark of Fig. 8 is shown plotted in Fig. 3.


Level of Mean Surface of Tin Coating
Level at Mean Surface of Steel Base
IN. X 10⁻³, NORMAL TO SURFACE.
70
60
50
40
30
20
10
2 1 0 1 2 3 4 5 6 7 8 9 10 11 12 13 14 15 16 17 18
IN. X 10⁻³, SUPERFICIALLY.

Specimen : Charcoal tinplate, average coating 0·00068 in. of tin on each side.
Number of complete fringes, 9.
Maximum depth, 0·000097 in.
Average axis superficially, 0·018 in.

Fig. 3.—Section through Grease Mark shown in Fig. 8.

Pores.

The positions of a number of normal pores were determined by a short application of the ferricyanide test paper.[2] As soon as a sufficient number of blue spots had appeared, the positions were marked off, the specimen was well washed in hot water and dried, and the pore sites were examined under the optical flat. A micrograph of the appearance of one of these pores is shown in Fig. 9. The section through the coating at the site of the pore shows the depth to be equal, within the limits of accuracy of the experiment, to the thickness of the tin coating, determined by the stripping method.[3]

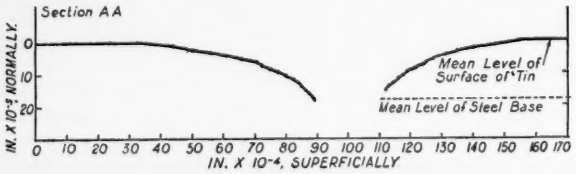

Section A A
Mean Level of Surface of Tin
Mean Level of Steel Base
IN. X 10⁻⁴ NORMALLY.
0
10
20
0 10 20 30 40 50 60 70 80 90 100 110 120 130 140 150 160 170
IN. X 10⁻⁴, SUPERFICIALLY

Fig. 4.—Examination of Normal Pore.

The dimensions of the pore, which can be appreciated from the section plotted in Fig. 4, are as follows :

Thickness of tin-coating	0·00018 in.
Depth of defect	0·00018(5) in. (17 fringes)
Diam. of affected area of coating	0·0125 in.
„ „ steel base exposure	0·002 in.

Method of Imposed Fringes.

For the examination of surface effects such as crystal boundaries and grease lines, it was found convenient to introduce a slight modification of the method.

In the modified method a set of fringes is imposed on the field by tilting the optical flat with respect to the specimen. This is quite conveniently arranged by inserting a piece of thin material between the flat and the specimen at one side, or by imparting a difference of tightness between the thumbscrews at either end of the jig. (Fig. 5.)

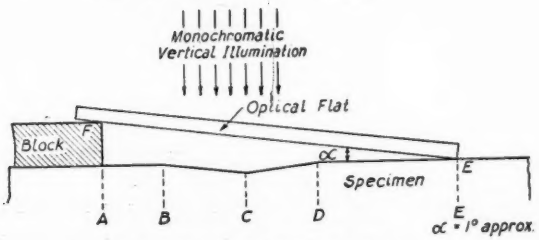


FIG. 5.—Diagram of Modified Method.

The system illustrated in the figure would give a normal spacing of fringes of, say, n per in. (*i.e.*, along AB and DE). Along BC, n decreases to zero as BC approaches a direction parallel to EF. Along CD, n increases until the fringes are too close together to be resolved at the magnification in use.

In this method, although the base-line or " sea-level " of the contours is removed from the horizontal, the fringes can still be considered as lines of iso-separation between the surface points of the specimen and the optical flat. A suitable angle between the optical flat and the specimen was found to be of the order of 1°, giving a fringe spacing of 1,620 per in. By the use of this method it will be seen that it is possible to appreciate more accurately the position of maximum depth of an irregularity having the shape of a narrow channel. (Fig. 6.)

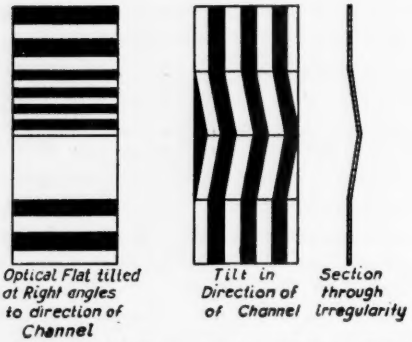

Optical Flat tilted at Right angles to direction of Channel
Tilt in Direction of of Channel
Section through Irregularity

FIG. 6.—Diagrammatic Representation of Fringes, Method 2.

A tin surface crystal boundary (unetched) examined in this way is illustrated in the photomicrograph, Fig. 10; the section is plotted in Fig. 7.

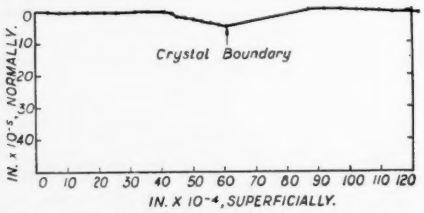


FIG. 7.—Section of Tin Surface Crystal Boundary depicted in Fig. 10.

It is interesting to note in this respect that the effect of the contraction at the boundary occurs over a comparatively wide area. The width of the boundary was, by ordinary illumination, apparently of the order of 0·0002 in., while the actual boundary effect existed over a band nearly 0 005 in. wide. The maximum depth of the boundary channel was approximately 0·00005 in.

Summary of Irregularities Examined.

Four grades of tinplates were examined, *viz.* :

Ordinary coke tinplate . .	1 lb. 12 oz. (coating thickness 0·000105 in.).
,, charcoal tinplate .	2 lb. 4 oz. (coating thickness 0·000135 in.).
Charcoal tinplate . . .	3 lb. 0 oz. (coating thickness 0·00018 in.).
,, ,, . . .	11 lb. 5 oz. (coating thickness 0·00068 in.).

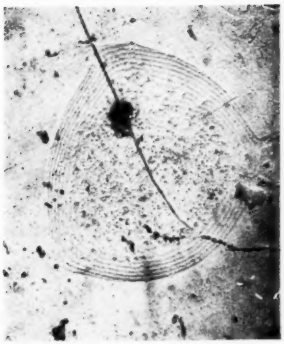


FIG. 8.—Grease Mark. × 80.

FIG. 9.—Normal Pore. × 80.

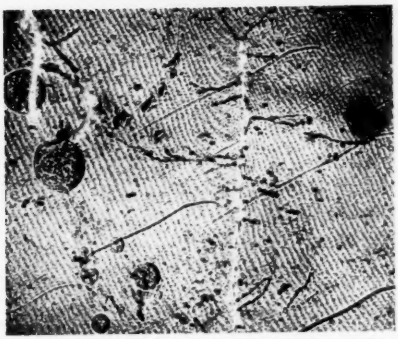


FIG. 10.—Tin Surface Crystal Boundary. × 80.

[*Hoare & Chalmers.*

To face p. 140.

Pores.—Normal pores, located by the ferricyanide test, were found to be hollows in the tin coating equal in depth to the thickness of the coating.

Crystal Boundaries.—Unetched tin crystal boundaries were found to be grooves or channels in the surface. In the case of the 3 lb. grade the maximum depth was approximately 0·00005 in.

Ripple Marks.—Ripple marks (at right angles to the direction of tinning), were found to be parallel ridges and valleys in the tin coating, having an undulating or wavelike contour in the direction of tinning. In a specimen examined, the half-depth of the valleys was approximately one-third of the mean coating thickness.

Grease Marks.—Isolated grease marks were found to be roughly circular hollows in the coating varying greatly in depth even in the same specimen. In no case did the depth approximate to the coating thickness, except where there was evidence that the mark was coincident with a normal pore. Excluding the 11 lb. grade charcoal tinplate, a fair percentage of the grease marks examined were of sufficient depth to penetrate to the compound layer. In the 11 lb. grade the concavities were of a depth generally not more than 25 per cent. of the coating thickness.

Grease Lines.—Grease lines were found to be narrow channels in the coating. At the present stage it has not been possible to measure the depth accurately since the gradient appears to be exceedingly steep. Some of the lines examined, however, appear to be sufficiently deep to penetrate to the compound layer.

Comments on the Method.

The authors wish to emphasise that the present paper is intended as a preliminary description of the method, and it is hoped to extend the application of the technique into fields of higher magnifications and other surfaces.

It should be noted that the successful application of the method does not depend on the possibility of the fringes being photographed. In practice, it is quite sufficient to plot the section by means of a mechanical stage, or an ocular scale ; or where the depth only is required, merely to count the number of complete fringes, each fringe representing a depth of approximately 0·00001 in.

Some difficulty was experienced originally in obtaining a sufficiently bright image for photography, and in this respect the use of Ilford " Double Xpres " plates was a refinement. These plates have a sensitivity peak at 560$\mu\mu$ and are thus eminently suitable for the particular light in use.

The method would appear to be applicable to the study of the surface contours of hot-dipped, electrodeposited and other coatings.

It is possible that the technique may also be of value in investigating some of the intrinsic properties of metals, &c., *e.g.*, twinning, the formation of etching pits, slip-bands, &c.

The authors wish to thank the International Tin Research and Development Council for permission to publish this work, and for the grant which enabled the work to be carried out. Thanks are also due to the Governors of the Sir John Cass Technical Institute and the Principal, Mr. G. Patchin, A.R.S.M., M.I.M.M., for facilities; to Mr. J. C. Prytherch, B.Sc., who did most of the stripping tests; and to Mr. D. J. Macnaughtan, F.Inst.P., for his advice and interest.

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(1) W. E. Hoare: *Journal of the Iron and Steel Institute*, 1934, No. I. p. 253.
(2) D. J. Macnaughtan, S. G. Clarke and J. C. Prytherch: *Journal of the Iron and Steel Institute*, 1932, No. I. p. 159.
(3) S. G. Clarke: *Analyst*, 1934, vol. 59, p. 525.

SOME ASPECTS OF THE FATIGUE PROPERTIES OF PATENTED STEEL WIRES.

II.—NOTE ON THE EFFECT OF LOW-TEMPERATURE HEAT TREATMENT.*

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(Musselburgh.)

Abstract.

Low-temperature tempering of cold-drawn wire is an essential process when the material is used in the form of a spring. The effect of this treatment on the fatigue properties has been examined for four steels of 0·37, 0·55, 0·79 and 0·86 per cent. carbon content respectively, drawn to varying reductions, from 25 to 90 per cent., and tempered in the range 150° C. to 400° C. The wires were free from decarburisation but not polished. The following points were noted:

Generally there is an increase in the limiting fatigue strength, the optimum temperature being of the order of 200° C. (for one steel, 0·79 per cent. carbon, it was 150° C.).

With all the steels examined, there is a critical reduction, at which point under certain tempering conditions the limiting fatigue stress may fall considerably below that for the as-drawn condition. This critical reduction becomes progressively lower as the carbon content increases.

The so-called cleanliness of the steel seems to have a bearing upon the results. The steel with the least amount of manganese, sulphur, and phosphorus present gave the most widely varying figures, while for the steel containing the most of these elements the fatigue limit never fell below that for the as-drawn condition. The results generally support the precipitation theory of hardening.

Fatigue limits as high as ± 44·9 tons per sq. in. were obtained, and it is suggested that if the materials are polished, a fatigue limit of the order of ± 55 tons per sq. in. should be possible

As far as has been ascertained there is no direct relationship between the usual physical tests (bends, torsions, &c.) and the limiting fatigue stress.

The effect of low temperature heat treatment upon the properties of cold-worked materials has been studied by many investigators,[1-16] and it has been found that tempering at temperatures up to 400° C. after cold-working modifies the properties to a considerable degree, and advantage is taken of this in many branches of industry. The tensile strength and hardness usually increase up to a certain point, and the same applies to the elonga-

* Received June 4, 1935. The research reported in this paper was carried out with the aid of a grant from the Carnegie Research Fund of the Iron and Steel Institute.

tion, yield point, and elastic limit, while it is usual for the ductility, as shown by the bend and torsion tests, to decrease. It is generally suggested that the optimum temperature for this type of heat treatment lies between 200° C. and 250° C. for plain carbon steels, and it is interesting to note that Pomp and Knackstedt[12] suggested some years ago that to draw steel wire at temperatures between 200° C. and 300° C. might produce highly elastic material.

However, very little investigation seems to have been made upon the modification of the fatigue properties of cold drawn steel wire after such heat treatment. It has been shown by Merrils[4] in the case of Armco iron that tempering at 375° C. after drawing has very little, if any, effect upon the limiting fatigue stress, but that the endurance properties at stresses above the limiting fatigue stress are considerably less than in the as-drawn condition, and the same applies to cold-rolled streamline wires of 0·49 per cent. carbon content. Similar results are reported by Adam[15] for a steel of 0·45 per cent. carbon content, and the general opinion expressed by these two investigators, from the results of their experiments, is that the tempering of cold worked steels is harmful because it impairs their resistance to overloading, a condition which often occurs in practice. But it should be noted that the tempering temperature used by both Merrils and Adam was 375° C., and higher than the suggested optimum temperature which is indicated above. Lea and Batey[16] report an increase in the torsional fatigue limit of cold-drawn steel wire after heat treatment at 300° C.

The work reported in this paper was originally contemplated in order to obtain some idea of the effect of tempering conditions upon the fatigue properties of wires used for the manufacture of helical springs, but seeing that rather extraordinary results were obtained in preliminary tests on the first steel examined (No. 3, 0·79 per cent. carbon), it was decided to extend the investigation to other steels, and tests have been carried out covering a range of tempering from 150° C. to 400° C. and reduction by drawing varying from 25 to 91 per cent. for four steels of varying carbon content. In addition to fatigue tests, results of further physical tests are reported ; these were made to ascertain if there is any correlation between the fatigue results and general physical properties of wires after tempering.

Preparation of Material.

Initially it was intended to confine the investigation to three steels differing fairly widely in carbon content. These steels, of 0·79, 0·55 and 0·37 per cent. carbon content, were examined in this respective order, and as the work progressed it became obvious from the results that data on a still higher carbon steel

would be of distinct interest, so a fourth steel of 0·86 per cent. carbon content was included.

TABLE I.—*Analyses of Steels Used.*

Steel No.	Carbon. %	Silicon. %	Manganese. %	Sulphur. %	Phosphorus. %
1	0·37	0·12	0·63	0·026	0·024
2	0·55	0·16	0·79	0·029	0·033
3	0·79	0·25	0·49	0·011	0·024
4	0·86	0·20	0·65	0·030	0·028

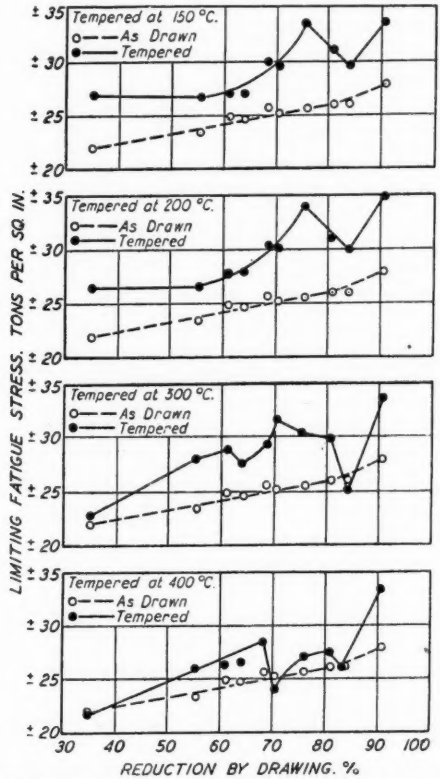

FIG. 1.—0·37% Carbon Steel Wires, free from decarburisation. Limiting fatigue stress after tempering.

The complete analyses of the four steels are given in Table I. In each case the test wire was drawn from lead-cooled patented rod free from decarburisation. The decarburised layer was removed by grinding and complete freedom from surface decar-

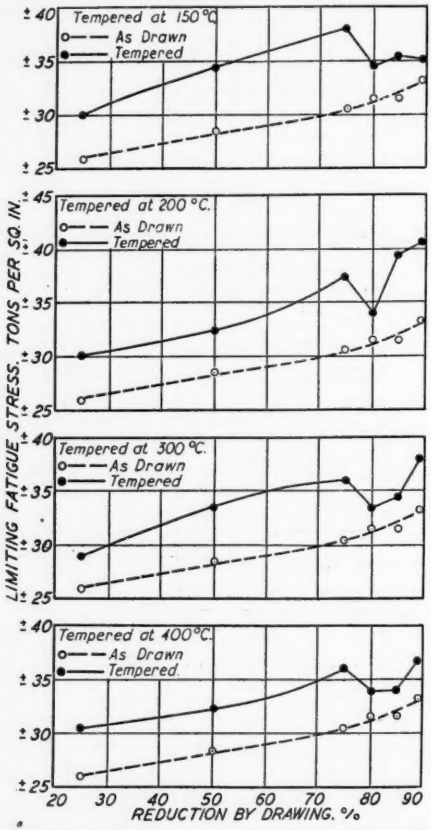

FIG. 2.—0·55% Carbon Steel Wires, free from decarburisation. Limiting fatigue stress after tempering.

burisation ensured by careful microscopic examination. A complete series of samples was prepared to represent fairly comprehensively the complete range of cold working to which such wire is normally submitted. The percentage reductions by drawing varied somewhat with different steels; the actual values are

given in the accompanying tables. For the convenience of testing, all samples were drawn to finish at a size of 0·080 in. in

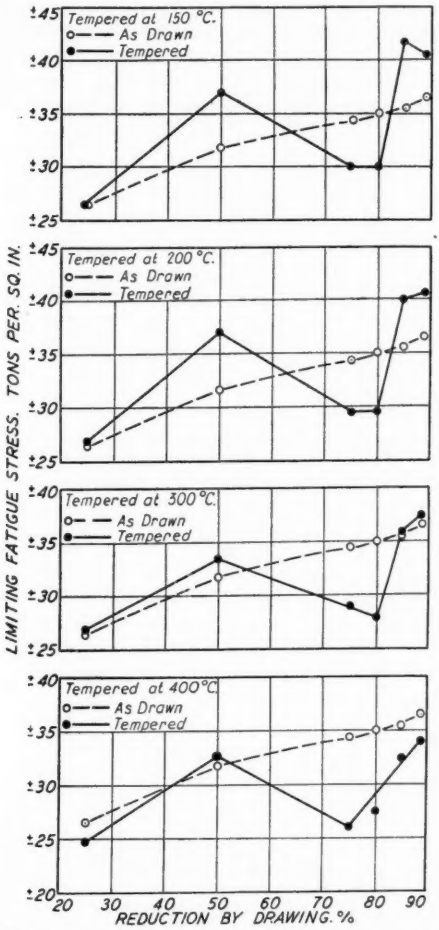

FIG. 3.—0·79% Carbon Steel Wires, free from decarburisation. Limiting fatigue stress after tempering.

diam. except for one series with steel No. 3. It will be appreciated that in the case of the wire drawn to 0·080 in. in diam., the size

of the commencing patented rod varied with the percentage reduction required, but in the special case the test samples were all drawn from the same size of patented rod and taken off at various stages in the drawing, in order to obtain some information on the effect of size upon the fatigue limit value.

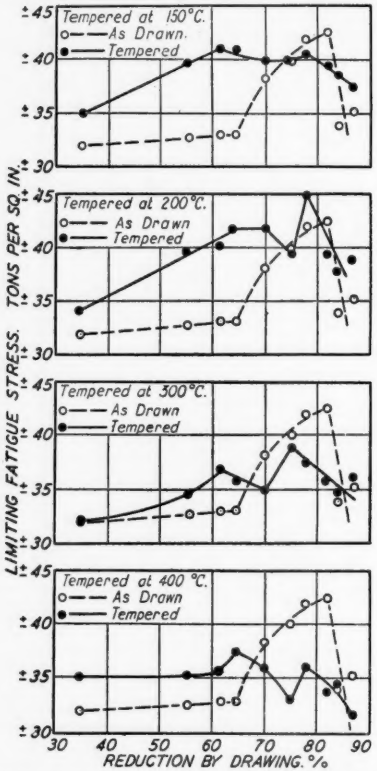

FIG. 4.—0·86% Carbon Steel Wires, free from decarburisation. Limiting fatigue stress after tempering.

The patenting operations were carried out within the temperature range 1,000° to 1,020° C., and the lead bath temperature for quenching was maintained in the range 480° to 500° C.

The cold-drawn test samples were tempered at four temperatures, namely, 150°, 200°, 300° and 400° C. In every case the samples were maintained at the particular temperature for 20

min. The heat treatment was carried out in an oil bath for the lower temperatures, namely 150° C. and 200° C., and in a salt bath for the remaining two.

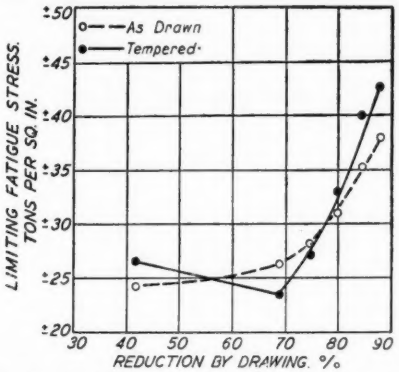

FIG. 5.—0·79% Carbon Steel Wires, free from decarburisation. Limiting fatigue stress after tempering, taken at various stages in reduction.

TABLE II.—*Mechanical Properties of* 0·37 *per Cent. Carbon, Patented Wires in As-Drawn Condition.*

0·37 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires in as-drawn condition.

Reduction by Drawing. %	No. of Holes.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
35·0	1	65·9	20	32	± 22·2	0·675
55·5	2	69·0	21	35	± 23·5	0·681
61·0	3	73·7	27	36	± 24·9	0·675
64·0	3	73·9	25	36	± 24·7	0·669
68·0	3	74·5	24	41	± 26·0	0·699
70·5	4	78·2	24	40	± 25·3	0·647
76·8	4	82·2	24	41	± 25·9	0·631
81·0	5	84·0	20	41	± 26·2	0·624
84·0	6	88·2	20	37	± 26·2	0·594
91·0	8	101·0	20	34	± 28·2	0·556

Fatigue tests, using the Haigh-Robertson fatigue machine, were made on the complete series of tempered wires and also on the same wires in the as-drawn condition.

Other physical tests carried out included the bend test (180° bends over 7½ mm. radius jaws) torsions on a gauge length of

100 times the diameter, tensile strength, and elastic limit, or limit of proportionality. In the latter determination the modulus of elasticity was also ascertained, and it was found that the variations in the value were insufficient to affect the fatigue calculations.

The complete physical test results are given in the accompanying tables.

TABLE III.—*Mechanical Properties of* 0·37 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·37 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 150° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
35·0	32·3	64·4	22	10	± 27·0	0·841
55·5	34·1	66·7	20	36	± 26·7	0·802
61·0	37·7	69·8	22	38	± 27·1	0·775
64·0	39·5	70·6	23	41	± 27·1	0·778
68·0	38·5	71·9	22	36	± 30·0	0·835
70·5	36·8	75·5	24	38	± 29·5	0·782
76·8	46·7	78·2	24	38	± 33·6	0·860
81·0	43·0	80·6	22	40	± 30·9	0·765
84·0	47·5	85·4	17	32	± 29·7	0·697
91·0	57·4	95·5	17	35	± 33·6	0·694

TABLE IV.—*Mechanical Properties of* 0·37 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·37 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 200° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
35·0	35·9	65·0	18	10	± 26·6	0·820
55·5	37·7	67·0	19	41	± 26·7	0·798
61·0	43·9	70·5	22	35	± 27·7	0·786
64·0	40·4	70·5	16	35	± 27·8	0·789
68·0	43·0	75·0	22	39	± 30·3	0·810
70·5	46·7	76·0	22	35	± 30·1	0·793
76·8	52·0	79·5	24	43	± 34·0	0·856
81·0	53·8	80·3	20	38	± 31·1	0·774
84·0	50·2	87·6	19	35	± 28·9	0·661
91·0	62·8	97·6	20	33	± 35·2	0·722

TABLE V.—*Mechanical Properties of* 0·37 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·37 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 300° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
35·0	39·5	62·6	23	9	± 23·0	0·743
55·5	39·5	63·8	20	33	± 28·0	0·894
61·0	42·1	67·1	21	28	± 28·7	0·857
64·0	50·2	67·9	24	28	± 27·4	0·808
68·0	49·4	72·0	23	35	± 29·0	0·806
70·5	46·7	73·5	22	31	± 31·6	0·860
76·8	52·0	77·7	23	34	± 30·5	0·785
81·0	57·4	79·5	22	35	± 29·6	0·744
84·0	66·4	83·0	18	24	± 25·2	0·607
91·0	68·1	91·0	18	35	± 33·7	0·730

TABLE VI.—*Mechanical Properties of* 0·37 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·37 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 400° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
35·0	39·5	64·2	20	9	± 21·7	0·675
55·5	47·5	67·0	21	34	± 26·0	0·765
61·0	46·7	67·3	23	29	± 26·3	0·781
64·0	49·4	68·0	21	30	± 26·7	0·785
68·0	50·2	72·5	21	27	± 28·4	0·768
70·5	55·6	74·4	20	30	± 24·0	0·646
76·8	55·6	75·5	20	30	± 27·2	0·720
81·0	55·6	77·6	20	36	± 27·6	0·711
84·0	60·1	81·5	17	34	± 26·0	0·638
91·0	70·9	86·9	18	34	± 33·5	0·771

DISCUSSION OF RESULTS.

The data obtained are given in Tables II. to XXIII. and the limiting fatigue stresses are shown graphically in Figs. 1 to 5,

being plotted against the percentage reduction by drawing for each of the tempering temperatures.

The following points were noted :

(1) Usually there is an increase in the limiting fatigue strength after tempering up to approximately 60 per cent. reduction by

TABLE VII.—*Mechanical Properties of 0·55 per Cent. Carbon, Patented Wires in As-Drawn Condition.*

0·55 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires in as-drawn condition.

Reduction by Drawing. %	No. of Holes.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	1	84·5	14	36	± 26·4	0·629
50	2	93·5	18	36	± 28·7	0·614
75	4	103·1	22	38	± 30·8	0·594
80	5	105·5	24	39	± 32·0	0·605
85	6	117·5	21	43	± 32·6	0·555
89	7	122·0	18	36	± 33·4	0·547

TABLE VIII.—*Mechanical Properties of 0·55 per Cent. Carbon, Patented Wires in Tempered Condition.*

0·55 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 150° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	48·4	80·1	13	36	± 30·1	0·755
50	53·8	89·1	18	35	± 34·4	0·772
75	57·4	102·7	23	30	± 37·8	0·735
80	53·8	109·0	24	38	± 34·6	0·635
85	82·5	120·1	23	42	± 35·6	0·594
89	80·7	123·1	16	34	± 35·2	0·570

drawing, but beyond this point a change takes place. In all cases there is a definite break in the curves and in many cases the limiting fatigue stress falls below the value for the as-drawn condition ; with those which do not fall below this value the

limiting fatigue stress is only slightly raised in comparison with the other percentage reductions.

(2) The position of the discontinuity noted in (1) varies with the carbon content, becoming progressively lower as the carbon

TABLE IX.—*Mechanical Properties of 0·55 per Cent. Carbon, Patented Wires in Tempered Condition.*

0·55 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 200° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	59·2	82·0	13	21	± 30·2	0·735
50	68·1	97·1	17	20	± 32·3	0·666
75	73·5	99·0	24	38	± 37·3	0·754
80	72·6	105·5	24	30	± 34·0	0·645
85	75·4	118·9	21	37	± 39·4	0·677
89	66·8	127·3	16	31	± 40·5	0·635

TABLE X.—*Mechanical Properties of 0·55 per Cent. Carbon, Patented Wires in Tempered Condition.*

0·55 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 300° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	53·8	81·0	14	4	± 28·4	0·701
50	62·8	86·4	19	7	± 33·5	0·775
75	71·8	100·9	24	33	± 36·1	0·715
80	77·3	98·2	23	41	± 33·6	0·685
85	75·4	109·9	20	32	± 34·2	0·622
89	71·8	103·5	15	30	± 38·0	0·695

content increases. Thus with steel No. 1, containing 0·37 per cent. of carbon, it occurs at about 84 per cent. reduction, although when the tempering temperature is above 300° C. there are signs of the formation of a second break in the curve; with steel

No. 2, containing 0·55 per cent. of carbon, the minimum value is reached at 80 per cent. reduction, and with steel No. 3 (0·79 per cent. carbon) it occurs at 75 per cent. to 80 per cent. reduction, and with steel No. 4 (0·86 per cent. carbon)

TABLE XI.—*Mechanical Properties of* 0·55 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·55 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 400° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	50·2	73·8	15	6	± 30·5	0·827
50	57·4	79·5	16	9	± 32·4	0·815
75	68·8	87·4	16	30	± 36·0	0·824
80	76·3	86·4	17	33	± 34·0	0·788
85	71·8	91·8	13	32	± 34·2	0·745
89	74·5	93·4	9	27	± 36·5	0·780

TABLE XII.—*Mechanical Properties of* 0·79 *per Cent. Carbon, Patented Wires in As-Drawn Condition.*

0·79 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires in as-drawn condition.

Reduction by Drawing. %	No. of Holes.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	1	98·2	15	16	± 26·8	0·546
50	2	114·8	19	25	± 31·8	0·554
75	4	133·0	22	41	± 34·8	0·524
80	5	137·0	22	43	± 35·0	0·511
85	6	138·0	21	29	± 35·7	0·513
89	8	141·0	21	24	± 36·2	0·510

at 70 to 75 per cent. reduction. In the latter steel there are distinct signs that it is overdrawn after a reduction of 82 per cent., because the limiting fatigue strength in the as-drawn condition decreases after this point and there is very little recovery after tempering.

(3) High values of the limiting fatigue stress may be obtained under certain conditions. With steel No. 1, the maximum value obtained was ± 35 tons per sq. in. when drawn to 91 per cent. reduction and tempered at 200° C. Steel No. 2 gives a maximum

TABLE XIII.—*Mechanical Properties of* 0·79 *per Cent. Carbon. Patented Wires in Tempered Condition.*

0·79 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 150° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	63·0	103·0	31	32	± 26·5	0·515
50	67·2	111·8	20	18	± 36·8	0·660
75	70·0	123·9	32	37	± 30·0	0·485
80	80·7	129·5	32	37	± 30·0	0·463
85	80·7	131·0	21	30	± 41·5	0·634
89	89·8	136·8	21	24	± 40·3	0·590

TABLE XIV.—*Mechanical Properties of* 0·79 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·79 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 200° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	62·7	102·6	32	33	± 27·0	0·526
50	67·3	112·0	19	19	± 37·0	0·660
75	80·7	126·1	30	35	± 29·5	0·468
80	77·0	129·1	32	35	± 29·5	0·454
85	85·2	129·9	21	31	± 40·0	0·616
89	77·0	140·2	21	25	± 40·5	0·578

value of ± 40·5 tons per sq. in. when drawn to 89 per cent. reduction and tempered at 200° C., and steel No. 3, when drawn to 85 per cent. reduction and tempered at 150° C., gives a maximum value of ± 41·4 tons per sq. in. In the case of steel

No. 4, a maximum of ± 44·9 tons per sq. in. is obtained with 78 per cent. reduction after tempering at 200° C.

When it is pointed out that none of the specimens tested was in a polished condition—that is, the wires would still contain

TABLE XV.—*Mechanical Properties of* 0·79 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·79 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 300° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	44·7	96·9	29	16	± 26·5	0·558
50	70·0	104·1	19	12	± 33·5	0·645
75	89·7	118·3	32	26	± 29·0	0·514
80	64·5	122·0	31	25	± 28·0	0·458
85	85·0	124·2	21	27	± 36·0	0·580
89	80·6	129·5	21	28	± 37·5	0·580

TABLE XVI.—*Mechanical Properties of* 0·79 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·79 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 400° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
25	53·8	85·6	22	19	± 24·8	0·580
50	62·7	91·0	12	9	± 32·5	0·714
75	62·7	98·6	19	17	± 26·0	0·528
80	71·5	101·2	16	20	± 27·5	0·542
85	75·5	104·1	8	22	± 32·5	0·624
89	71·5	106·1	10	24	± 34·0	0·640

slight draw marks and defects—it seems rather amazing that such high values should be obtained; and since polishing can result in an increase of about 24 per cent. in the limiting fatigue stress, as the authors[17] have already shown, the true fatigue

limits of these materials ought to be of the order of from ±47 to ± 55 tons per sq. in. This would give ratios of limiting fatigue stress (total range) to tensile stress of the order of 0·80 in the case of the higher carbon content, rising to unity with the lower carbon

TABLE XVII.—*Mechanical Properties of* 0·86 *per Cent. Carbon, Patented Wires in As-Drawn Condition.*

0·86 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires in as-drawn condition.

Reduction by Drawing. %	No. of Holes.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress. (Total Range) to Tensile Strength.
35·0	1	112·0	15	19	± 31·8	0·568
55·5	2	126·8	20	26	± 32·7	0·516
61·5	3	128·0	20	31	± 33·0	0·516
64·5	3	130·2	20	23	± 33·0	0·506
70·0	4	137·3	18	24	± 37·4	0·544
75·0	4	140·5	18	25	± 40·1	0·570
78·0	5	145·0	18	25	± 42·1	0·581
82·0	5	156·3	14	23	± 42·4	0·541
84·0	6	165·2	16	22	± 34·2	0·413
87·0	7	165·0	12	21	± 35·4	0·429

TABLE XVIII.—*Mechanical Properties of* 0·86 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·86 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 150° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
35·0	69·5	107·8	15	6	± 35·1	0·652
55·5	67·3	120·3	19	18	± 40·4	0·671
61·5	62·8	124·0	19	23	± 41·0	0·660
64·5	83·0	125·7	20	31	± 40·9	0·650
70 0	83·0	129·5	18	33	± 39·9	0·616
75·0	80·8	138·0	18	27	± 40·0	0·578
78·0	80·8	147·2	16	20	± 40·6	0·550
82·0	92·0	152·5	15	22	± 39·5	0·518
84·0	89·7	165·0	13	21	± 38·7	0·469
87·0	89·7	166·0	10	19	± 36·5	0·440

content, values which to the authors' knowledge have not been obtained with cold-worked materials except mild steel.

(4) Generally the maximum increases in the limiting fatigue stresses for all percentage reductions occur after tempering at

TABLE XIX.—*Mechanical Properties of* 0·86 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·86 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 200° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
35·0	78·5	108·4	14	5	± 34·3	0·596
55·5	78·5	119·5	19	23	± 40·5	0·678
61·5	78·5	124·7	19	17	± 40·3	0·648
64·5	96·5	124·9	19	22	± 41·6	0·665
70·0	96·5	132·6	17	29	± 41·6	0·627
75·0	98·8	134·1	20	23	± 39·0	0·582
78·0	101·0	141·7	17	29	± 45·0	0·636
82·0	112·2	151·0	14	27	± 39·5	0·523
84·0	130·0	157·0	13	22	± 37·0	0·472
87·0	125·7	161·1	10	20	± 39·3	0·488

TABLE XX.—*Mechanical Properties of* 0·86 *per Cent. Carbon, Patented Wires in Tempered Condition.*

0·86 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 300° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
35·0	74·0	106·8	14	4	± 31·8	0·579
55·5	83·0	103·3	17	28	± 34·6	0·596
61·5	87·5	116·9	18	25	± 36·8	0·630
64·5	83·0	117·4	19	29	± 36·1	0·614
70·0	87·5	124·0	17	25	± 35·0	0·564
75·0	89·7	126·2	19	28	± 39·0	0·617
78·0	89·7	130·9	17	22	± 37·6	0·575
82·0	107·8	139·0	16	22	± 36·1	0·519
84·0	112·2	148·0	12	26	± 34·8	0·470
87·0	112·2	148·0	12	24	± 36·4	0·491

200° C. except in the case of steel No. 3, containing 0·79 per cent. of carbon, where it is possible that 150° C. gives the best results. This is quite in accordance with the remark made in the intro-

TABLE XXI.—*Mechanical Properties of 0·86 per Cent. Carbon, Patented Wires in Tempered Condition.*

0·86 per cent. carbon lead-cooled patented wires drawn from rods free from decarburisation. Wires tempered 20 min. at 400° C.

Reduction by Drawing. %	Elastic Limit. Tons per sq. in.	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress (Total Range) to Tensile Strength.
35·0	67·3	94·2	10	7	± 35·1	0·745
55·5	71·8	98·5	8	6	± 35·4	0·718
61·5	71·8	105·0	9	7	± 35·6	0·678
64·5	74·0	105·8	8	4	± 37·5	0·710
70·5	76·3	105·0	8	25	± 36·0	0·684
75·0	80·8	107·4	8	27	± 33·0	0·614
78·0	85·1	111·2	7	17	± 36·1	0·646
82·0	87·5	114·8	7	16	± 33·9	0·592
84·0	89·7	119·2	4	14	± 34·5	0·577
87·0	94·3	121·0	4	14	± 31·7	0·524

TABLE XXII.—*Mechanical Properties of 0·79 per Cent. Carbon, Patented Wires in As-Drawn Condition.*

0·79 per cent. carbon lead-cooled patented wires drawn from rod 0·232 in. diam., free from decarburisation. Wires in as-drawn condition.

Diameter of Wire. In.	Reduction by Drawing. %	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress. (Total Range) to Tensile Strength.
0·144	41·5	100·3	9	31	± 24·6	0·488
0·128	69·5	103·0	12	33	± 26·3	0·511
0·116	75·0	103·6	13	34	± 28·3	0·521
0·104	80·0	115·6	15	31	± 30·9	0·534
0·092	84·2	124·5	20	33	± 35·4	0·572
0·080	88·0	143·0	25	27	± 38·1	0·532

duction, that the optimum temperature for tempering cold-worked plain carbon steels lies between 200° C. and 250° C.

Special reference ought to be made to Fig. 5, which gives results for 0·79 per cent. carbon material tested at varying stages

in reduction instead of all at one size. Comparison with Fig. 3 will show that the fatigue ranges are not quite as high as when all are drawn to 0·080 in. in diam., except for the highest percentage reductions where the sizes are comparable. Hence it is concluded that there is a "size effect," which can probably be traced to the patenting process, as the larger size of rod tends to cool more slowly, thereby causing slight differences in the sorbitic structure which modify the properties of the drawn wire. In this curve the discontinuity after heat treatment is confirmed, but it occurs at a lower percentage reduction, 70 per cent. instead

TABLE XXIII.—*Mechanical Properties of 0·79 per Cent. Carbon, Patented Wires in Tempered Condition.*

0·79 per cent. carbon lead-cooled patented wires drawn from rod 0·232 in. in diam., free from decarburisation. Wires tempered 30 min. at 200° C.

Diameter of Wire. In.	Reduction by Drawing. %	Tensile Strength. Tons per sq. in.	Bends (Average) Angle 180° Radius 7·5 mm.	Torsions (Average) Gauge Length = Diam. × 100.	Limiting Fatigue Stress. Tons per sq. in.	Ratio of Limiting Fatigue Stress. (Total Range) to Tensile Strength.
0·144	41·5	103·0	8	30	± 26·4	0·514
0·128	69·5	106·5	9	32	± 23·7	0·445
0·116	75·0	110·1	13	31	± 27·4	0·496
0·104	80·0	120·5	15	31	± 33·0	0·547
0·092	84·2	128·1	20	32	± 40·1	0·626
0·080	88·0	140·0	27	25	± 42·8	0·612

of 75 to 80 per cent. Also it should be pointed out that the time of tempering was longer; this may play some part, but until it has been investigated further no conclusions can be made. However, it does tend to show that the observations made above may have to be modified in degree when applying them to larger sized materials, and all that they show is the tendency rather than the true condition.

The authors still have under consideration the question of endurance at stresses above the limiting fatigue stress for these tempered materials. The analysis of the results obtained is not complete as yet and it is hoped to record them at some future date, but certain observations may be stated here. The endurance properties are related to the ductility of the material, and general inspection of the figures given in the tables shows that they are in accordance with previous results in that tempering causes a decrease in the bends and torsions. However, this is not always the case. It will be noted that it is usually marked for

the low and high percentage reductions, but with the intermediate reductions tempering up to 200° C. has very little effect when compared with the as-drawn condition, and in some cases there is an actual increase. This same effect applies to the endurance to repeated stresses; inspection of the fatigue curves shows an improvement compared with the as-drawn condition.

The question of the correlation of fatigue with other physical properties has received the attention of many investigators. It has been shown by the authors[17] that there is no apparent connection between the fatigue and other physical properties of wire except in the case of reverse bends, and this is only qualitative. This same statement applies to wires in the tempered condition when the results are considered as a whole.

The elastic limits, or more accurately limits of proportionality, have been determined, and general inspection of the results shows that tempering at 200° C. is sufficient to obtain the highest values; at higher temperatures there is little or no elevation. With steel No. 3 (0·79 per cent. carbon) the highest values can be obtained by tempering at 150° C. This is quite in accordance with some results published by Greaves,[6] who showed that cold-drawn wire must be heated to at least 200° C. before constant stress with constant strain can be established for the material at normal temperatures. Rees[11] also mentions a similar effect, namely, that the limit of proportionality increases up to 200° C., which is the optimum temperature. Hence it can be said that there is apparently no connection between the fatigue of tempered wires and the corresponding limit of proportionality, seeing that the limiting fatigue stress decreases while the limit of proportionality remains constant. Further, there seems to be no parallel discontinuity in the elastic limit as the percentage reduction increases, as compared with the limiting fatigue stress.

It has been considered by Gough[18] that the ultimate strength is the only mechanical property permitting correlation with alternate bending stress, and most investigators seem to concur with this opinion. The results of Forsman[19] show that high tensile strength is accompanied by high fatigue limit. However, this does not seem to be the case with cold-drawn wires, both in the as-drawn condition, and after tempering. It will be noted by reference to the tables that the ratios of the limiting fatigue stress to tensile stress for the wires which have been examined show a very wide variation. The maximum differences are: Steel No. 1 (0·37 per cent. carbon), 53 per cent.; steel No. 2 (0·55 per cent. carbon), 51 per cent.; steel No. 3 (0·79 per cent. carbon), 40 per cent.; and steel No. 4 (0·86 per cent. carbon), 74 per cent. These figures are calculated on the lowest observed values. Lessels[10] suggests that the true elastic limit, that is, the endurance limit, is related to the ultimate strength alone,

but adds that it is influenced by three factors, internal stress as measured by the yield stress, the elastic limit and the microstructure. It would seem from the authors' results that these factors are very variable, indeed so variable as to suggest that there is no practical relationship between these two properties for wire in the as-drawn and tempered conditions.

It is doubtful whether the true reason can be given for the results obtained and described in this paper. Several theories have been developed to explain the changes which take place in the properties of metals after cold-work, and these have been used to give reasons for the modifications produced by low-temperature heat treatment.

These theories and hypotheses can be placed under three headings: The amorphous metal theory, internal stress theories and precipitation theories. The amorphous metal theory, first suggested by Beilby, has been developed by Rosenhain and by Aitchison[3] for cold-worked materials. Grossmann and Snyder[10] offer a slight modification to make it applicable to the changes which take place after tempering cold-worked metals. They report that their evidence supports the hypothesis that these phenomena (changes in properties) originate in the amorphous metal produced by cold-work. The grains of the metal deform by block slip, the regions between becoming amorphous. The reheating causes a certain amount of rearrangement into thin layers which obstruct slip, but above a temperature of about 315° C. the size of the recrystallised layers has become too large to prevent slip.

It has been suggested by other investigators that conditions of internal stress, assumed to have been caused by atomic slip or glide in the crystal, due to deformation, are responsible for the change in the properties. This was used by Merrils[4] to explain why the fatigue and elastic limits of cold-worked materials could be raised by understressing with repeated stresses, and why the endurance properties were modified by heat treatment, it being assumed that a certain amount of relief of the internal stresses takes place. The possible variation of internal stress was used by Haigh[21] in the development of his hysteresis theory of fatigue.

The precipitation theory attacks the question from another point of view. It assumes that ageing, both normal and artificial, is caused by the precipitation of submicroscopic particles from the α-iron in the case of steels or the solvent metal in the case of other alloys. It is presumed that the precipitation causes distortion in the space lattice, thereby hindering slip. It is also considered that the precipitation is considerably modified by previous working. But, so far, neither the nature of the precipitate nor the mechanism of precipitation seem to have been established. Pfeil[22] suggests from his investigations that the pre-

cipitation of iron oxide from the crystal lattice of cold-drawn iron and low carbon steels due to the lessening solubility in cold-drawn material increases the tensile strength. Dean, Day and Gregg[13] conclude that commercial irons owe their property of hardening on reheating to 250° C. after cold-work to the precipitation of a small amount of iron nitride present. Mehl[23] states that it is impossible to say which element, carbon, oxygen, nitrogen or sulphur, is the cause of the phenomenon, and Sauveur[24] expresses a similar opinion. Walzel[25] noted that as the proportion of pig iron was reduced in the charge for making unkilled open-hearth steels, and as segregation was increased, the ageing tendency was likewise increased. French[26] goes further, and suggests that the particle deposition takes place in the slip bands, thereby obstructing slip and improving the resistance to fatigue.

It is difficult to assess the merits and demerits of the respective theories and suggestions, as there is considerable evidence to support all of them, and probably each one by itself is not sufficient to explain the changes which take place in the fatigue properties of cold-drawn wires after heat treatment. Most investigators seem to have confined their attention to wires which had been drawn to approximately 65 per cent. reduction and less, and very little is reported for higher percentage reduction-. It would seem that the discontinuity which is observed in all the curves in this paper for heat-treated wires, is due to some critical distortion which is influenced primarily by the carbon content, and it is interesting to note that a somewhat parallel effect was reported by Adam[15] in the case of the tensile strength; a definite break was noticed in the tensile-strength/percentage-reduction curve, and this occurs earlier as the carbon content increases. A similar effect was noted by the authors[17] for the endurance of cold-drawn wires, when in the region of 75 per cent. reduction either a maximum or a minimum in the endurance properties was obtained. But it is impossible to say whether the discontinuity in the limiting fatigue stress after tempering is due to the original internal stress conditions in the as-drawn material, either by themselves or because at this distortion precipitation is hindered; possibly the two factors are at work. However, it is interesting to examine steel No. 3 (0·79 per cent. of carbon). This gives the widest variations in the series examined. It is a Swedish steel, and inspection of Table I. will show that the manganese, sulphur, and phosphorus contents are considerably less than in the others, although the silicon is higher. Steel No. 2 (0·55 per cent. of carbon) gave the least variation in the fatigue curves. The limiting fatigue stress does not fall below that for the as-drawn condition with any of the tempering temperatures used. This steel carries the highest contents of manganese, sulphur and phosphorus in the series. Hence it would seem that

to a certain degree the results obtained can be modified considerably by the so-called impurities in steel, and this lends support to the precipitation theory. Bolsover[14] has stated that increasing phosphorus content has a particularly marked effect upon steels which have been reheated after cold working.

It is felt that the commercial aspect of these results is of importance, although, as stated above, they only show the tendency rather than the true condition. Before they can be applied commercially, a large amount of investigation will have to be done, but they indicate that if the maximum fatigue properties are required in helical springs and other articles made from cold-worked materials, careful attention will have to be paid to more factors than is realised at present. Chemical composition, amount of cold-working and subsequent heat treatment will all require attention, but it is felt that the enhanced qualities of the products would make the extra care worth while.

In conclusion, the authors wish to express their thanks to the directors of Bruntons (Musselburgh), Ltd., for permission granted and facilities given to carry out this work. The results given in this paper are part of a programme for which a grant has been received from the Carnegie Research Fund, and the authors express their indebtedness to the Council of the Iron and Steel Institute for this financial assistance.

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DISCUSSION.

Mr. H. SUTTON (Farnborough) said that the authors had placed on record the results of a very extensive investigation, and the data they had provided were of both practical and theoretical interest. At first sight the work would seem to be an essential asset to those engaged in the production of high-grade cold-drawn steel wires, as used for the manufacture of springs of high quality and reliability. The fact that such wires, free from surface decarburisation, were now manufactured on a commercial scale reflected very great credit upon the firms who had taken up the work and upon the metallurgists and other technicians responsible for the development of the necessary technical processes. It was, indeed, gratifying to learn that cold-drawing of the normalised carbon steel wires of 0·37 to 0·79 per cent. carbon content gave progressive improvement in fatigue properties, and that this seemed to obtain with increasing reductions of sectional area up to 90 per cent. It would interest him very much to know whether reductions of that kind—*i.e.*, up to 90 per cent. after normalising—represented the range which was of commercial importance, whether large reductions were ever likely to be wanted or whether the data given were what really mattered.

The data given and the results for the 0·79 per cent. carbon steel, for instance, would suggest that tempering for high fatigue properties must be conducted with very careful reference to the previous drawing operations to which the wire had been subjected, particularly when the drawing had not resulted in a very big reduction of section.

With reference to the " size effect," the lower fatigue properties of the wire drawn by moderate reductions from rod normalised at a diameter of 0·232 in., compared with those of wires normalised at diameters of rather more than 0·080 in. and then drawn down to 0·080 in., would seem, as the authors suggested, to be due to the different response to heat treatment in the two cases in view of the widely differing section. The longer tempering period employed in one series, Table XXIII., compared with the series of Table XIV., which related to the same steel, was presumably employed for some special reason.

The meaning of the percentage variation in fatigue ratios given at the bottom of p. 161 was not very clear to him. Personally, he found it rather more simple to study the ratios themselves. In the case of most of the materials which were giving high performance in engineering and transport machinery, the ratios of limiting fatigue stress to ultimate stress were fairly high.

It was pleasing to note that the authors were considering and working on the question of endurance at stresses higher than the fatigue limit. That matter was of considerable practical importance and was also of theoretical interest. The shape of the stress-cycle curves, the S/N curves, was of special interest, and he trusted that the authors would bear that in mind when they finally presented their work. The endurances for the limits given did not seem to be stated in the present paper. He assumed that the endurances were the same as in the previous paper, but it would be of value to put in the endurance on one or two of the diagrams for the benefit of new members.

Professor F. C. Thompson (Manchester) said the paper dealt with an aspect of the production of wire which had received far less attention than its importance merited. As a result, there were one or two experimental points concerning which it would be of interest if the authors could give information; the type of fatigue test used, for instance, was not mentioned. Other points which were really very important were the order of accuracy of the experiments and the reproducibility of the results. If very much weight were to be attached to kinks in curves, it was essential to know what the order of the experimental error was likely to be. It would also be interesting to know whether the authors had any evidence of the extent to which ageing took place after the tempering. They did not mention a fact which might be of very great importance, namely, the time at which the tests were made after the various mechanical and heat-treatment operations had been carried out. Finally, it would be of interest to all those concerned with the testing of cold-drawn wire if the authors would give some short description of the methods which they found most suitable for the determinations of the limit of proportionality and the modulus of elasticity.

Passing on from such experimental details to the matter of the paper, there were several very interesting points which had been raised. He was going to assume that ordinary ageing and the changes of properties which occurred on low-temperature heat treatment were—at any rate in their essentials—the same—an assumption which seemed to him to be justified, though it might, of course, not appear so to everyone. There was a growing body of evidence which indicated that the ageing process was not as simple as was at one time imagined. If one drew the curves for several ageing operations faithfully through the experimental points obtained, instead of getting the smooth curve which used to be, and was still, for that matter, generally believed to express the results, one obtained a curve with a series of small but quite definite periodic fluctuations. In wire drawing, fluctuations of a similar character had been obtained in work which Mr. Francis and he had published

on the influence of the die angle. He had no explanation whatever to offer for those fluctuations, but he was more and more convinced as time went on that they had a real existence. It was almost inevitable, therefore, that one should ask one's self whether the points which the authors had obtained in their curves might not be an indication of a similar type of fluctuation, in this particular case of an abnormally high magnitude. It would be of great interest if the authors could at some time study a small portion of one of those curves round the critical reductions, and, by obtaining a sufficient number of points near together, determine whether the abnormalities were not really due to a definite periodicity superimposed on a generally smooth curve.

There was no time to enter in detail into the very complex question of the explanation of these phenomena, but there was one aspect which ought certainly to be borne in mind, and which might be overlooked by those conversant only with the drawing of steel wire. If one drew any non-ferrous wire which did not belong to the alloy class that showed ageing properties, one did not in general find any indication of changes of properties on ageing. That seemed to put out of court any explanation based on an amorphous material, or on the presence of internal stresses. Amorphous material, if it were formed, and internal stresses should be present in those materials as much as in the steels. One was, therefore, left with the precipitation explanation as the only one which fitted the facts. The identity of the material which was precipitated was not yet established.

The authors mentioned the question of the modulus of elasticity and said that so far as their work was concerned they found little or no variation as a result of tempering. That was not in agreement with other results. He thought it had generally been found that the modulus of elasticity was definitely raised by the low-temperature heat treatment of cold-drawn wire, and he believed that there was some ground for the belief that one could possibly get a higher modulus of elasticity from a cold-worked and blued steel than from any other ferrous material.

CORRESPONDENCE.

Mr. E. A. Atkins (Warrington) wrote that the authors were to be commended for the very fine piece of research work that they had carried out and which should result, ultimately, in the production of wire to give the highest and most reliable results. It was noted that all the wires tested were ground to remove the decarburised layer and surface defects, and it was suggested that a further set of tests should be made on wire as commercially

produced, as these would show how far ordinary wire differed from the perfect product as used by the authors. The results should show users how far they could depend upon the quality of wires produced by the ordinary method of manufacture.

Dr. T. SWINDEN (Member of Council, Sheffield) wrote that the practice of reheating springs made from hard-drawn wire to approximately 200° C. was very old indeed, but it was only during the war when other problems directed attention to the changes in properties involved by this treatment that the matter received serious attention by way of organised investigation.

The work from Woolwich Arsenal by Greaves and Rees dealt with the matter in a comprehensive way with particular reference to gun-wrapping wire, and the same problem had since received extensive attention in connection with steel strip. These points were noted because of the general adoption of a temperature of the order of 200° C. as being the optimum temperature for treating high-carbon hard-drawn material.

The authors were to be congratulated on having worked out, in an admirable way, reliable data linking up low-temperature heat treatment with the percentage reduction in wire-drawing, and were to be complimented on a first-class piece of work.

It was noted that the tests were carried out on commercially drawn wire which, however, had been subjected during manufacture to the special treatment of having any decarburised surface removed, as specified for certain types of springs.

It was of importance to note that, if the anticipated still further improved results obtainable by polishing were achieved, the fatigue values would be of a very satisfactory order.

From the point of view of springs, no doubt those data were directly applicable to service, and that, one gathered, was the immediate objective of that piece of work.

It was of some importance, however, to remember that in other applications of hard-drawn wire, surface perfection was not possible, and, moreover, that during service, imperfections were developed in the wire. He had in mind particularly the development of small surface cracks arising in the martensitic layer often found in worn ropes, for example.

It would be of particular interest to know how the relationship between reduction in drawing and low-temperature treatment would work out, as interpreted by fatigue tests in the presence of surface imperfections of known dimensions. In brief, what was the order of notch sensitivity of wires produced from steels of a varying carbon content under the best conditions suggested by the authors?

Mr. A. B. WINTERBOTTOM (Trondheim, Norway) wrote that fibre structure change during heat treatment might be a factor worth considering along with those enumerated, and X-ray examination for this purpose might at the same time give information concerning surface stress. If the hoped-for 24 per cent. improvement on polishing was not substantially realised surface compressive stress effect might be indicated. It would be interesting to know whether the torsion and fatigue fractures exhibited any special features, *i.e.*, whether they were flat or helicoidal.

Professor F. C. LEA (Sheffield) wrote that the paper was of interest to him, as for some years he had been carrying out tests, in the time he was able to spare from the many other activities of a limited university laboratory, on the effect of cold work on the physical properties of metals. A summary of some tests which he and his co-workers carried out during the War, on the effect of varying tempering temperatures on cold-drawn tubes, was published by Professor C. F. Jenkins in "Materials for Aircraft." A paper dealing with the effect of tempering temperatures on the fatigue range of cold-drawn tubes was published in *Engineering*, 1927, vol. 124, Dec. 23, p. 797. That paper showed very clearly that the compressive strength of these tubes was raised by tempering at temperatures from 200° to 400° C., the temperature at which the maximum was reached depending to a certain extent upon the carbon content of the tubes. The experiments also showed that the limit of proportionality became a maximum when the tempering temperature was somewhere between 360° and 450° C. The effect of the tempering temperature on the fatigue range was quite marked, the maximum value being obtained at temperatures of the order of 360° C. In these tests the outer surface of the tube was ground away. A considerable amount of work had also been carried out on the effect of low temperature tempering on cold-drawn wires. Tests on certain wires, the results of which were published in a paper in the *Proceedings of the Institution of Mechanical Engineers*, 1928, vol. 2, p. 865, showed that the fatigue range could be raised 100 per cent. by heat treatment at 400° C. In other cases it was found that the fatigue range was only changed by a comparatively small amount. In some recent tests on a wire drawn to the D.T.D.5A. specification the following particulars were obtained :—

Analysis of Steel.

Carbon.	Manganese.	Silicon.	Sulphur.	Phosphorus.
%	%	%	%	%
0·78	0·69	0·13	0·035	0·034

Details of Drawing.

The initial rod size was $\frac{13}{32}$ in. in diam. This was ground to 0·370 in. in diam., then drawn to 0·160 in. diam. without further heat treatment.

Tensile Tests.

Reference No.	Description.	Ultimate Strength. Tons per sq. in.	Reduction of Area. %
..	As drawn	110·0	47·7
I.	As spun	109·1	56·7
II.	Heated to 250° C. . . .	114·7	52·7
III.	Heated to 350° C. . . .	114·4	33·8
IV.	Heated to 450° C. . . .	99·7	33·8

Tests on D.T.D.5A. Wire after "Blueing."

The specimens were tested in the alternating torsion machine with type metal ends cast on. The results are shown in Table A.

TABLE A.—*Repeated Stress Tests.*

Nature of Test	Range of Shear Stress. Tons per sq. in.	Repetitions of Stress.
(1) Tests on wire as spun, with no heat treatment.	± 21·1	Broke after 2,700,000
	± 20·6	„ „ 4,800,000
	± 20·2	„ „ 6,160,000
	± 19·8	Unbroken after 12,000,000
	± 18·5	„ „ 10,000,000
(2) Tests on wire after heating to 250° C.	± 21·3	Broke after 1,552,000
	± 20·6	„ „ 6,240,000
	± 20·3	„ „ 12,500,000
(3) Tests on wire after heating to 350° C.	± 21·9	Broke after 7,900,000
	± 21·2	„ „ 9,500,000
(4) Tests on wire after heating to 450° C.	± 21·6	Broke after 4,218,000
	± 20·0	Unbroken after 10,000,000

It would be seen from the table that the effect of heat treatment at various stresses up to 450° C. had very little effect upon the safe range of repeated stress. Figs. A and B showed the distribution of hardness in these wires. After final drawing the wire was spun to ensure straight lengths and then "blued" at the temperatures shown in the table. A great deal of work had yet to be done before it could finally be said what wires would have their fatigue ranges improved by blueing. The degree of improvement depended upon a number of circumstances and conditions. It was well-known that heating cold-drawn materials in the neighbourhood of 200° to 300° C. tended

to increase the brittleness of these materials, and for that reason the writer had generally thought it desirable to use tempering temperatures of the order of 400° C. It was, however, of considerable importance that that point should be further investigated, particularly in relationship to wires of various compositions and subjected to various degrees of cold work. It

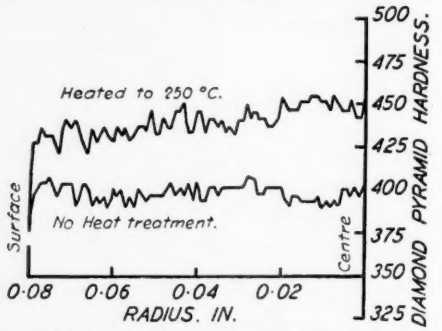

FIG. A.—Tests on D.T.D. 5A. Wire after "Blueing." The distribution of hardness in the wire with no heat treatment and after heating to 250 ° C.

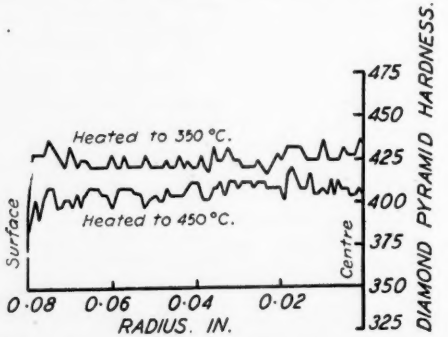

FIG. B.—Tests on D.T.D. 5A. Wire after "Blueing." The distribution of hardness in the wire after heating to 350° C. and 450° C.

would also be of interest if the effect of gas inclusions on the properties of the cold-drawn steel, especially with reference to fatigue ranges and heat-treatments, could be investigated. If they were of importance it would appear that although tending to increase the tensile strength they might introduce the possibility of intercrystalline, or crystal boundary, concentrations of stress which led to a lowering of the fatigue range. From

the results of the effect of occluded hydrogen[1] on the safe range of repeated stresses on low-carbon steels it seemed doubtful whether occluded gases were important; but it might be that they were important in cold-worked materials. The degree of crystal distortion and fracture at crystal boundaries produced by drawing might also be important. It would be of interest to investigate the effect of heat treatments under low-vacuum conditions.

AUTHORS' REPLY.

Dr. GOODACRE, in reply, said that Mr. Sutton asked whether the percentage reductions up to 90 per cent. came within the limits of practice. Actually, an old wire-drawer would always say that 75 per cent. reduction was the limit, though on occasion, it was possible to draw beyond this point. He wished to emphasise, however, that wire of ordinary grade was not at all the same as that with which the present paper was concerned, and that it was now known that if surface decarburisation was removed, the properties, compared with decarburised material, were enhanced.

The difference in the time of tempering to which Mr. Sutton referred was, unfortunately, accidental. The period intended was to have been 20 min. as for the smaller wires, but the material was left in the furnace for the longer period. He did not think that this would make much difference to the results, but this was only an opinion expressed without any evidence to support it.

On the question of the percentage variation of the ratio of the limiting fatigue stress to the tensile stress, this was given for each steel covering the whole series of percentage reductions and tempering temperatures. The variation was based upon the minimum ratio, and these figures were mentioned to show that it was impossible to forecast the fatigue limit from the ultimate strength value in the case of wire.

With regard to Professor Thompson's remarks, the absence of any reference to the fatigue test used was an unfortunate omission. For this investigation the new Haigh-Robertson fatigue testing machine was used, an alternate bending machine which had been developed for wire.[2] With regard to the order of accuracy and reproducibility of the results, the authors had tried this out time and time again, and they had found that the accuracy—or, rather, the reproducibility—was within $\pm 0{\cdot}5$ tons per sq. in. when estimating the limiting fatigue strength. Seeing

[1] F. C. Lea, "The Penetration of Hydrogen into Metal Cathodes and its effect upon the Tensile Properties of Metals and their Resistance to Repeated Stress." *Proceedings of the Royal Society*, 1929, A., vol. 123, pp. 171–185.

[2] This is now mentioned in the text.

that it was possible to be accurate within this limit, he did not think that experimental variation would affect their inferences from the curves in any way.

He confessed that the authors had not considered the possibilities of ageing after tempering, nor the lapse of time after heat treatment before the tests were made. The usual procedure in the preparation of the material was as follows : All the wires of one steel were tempered on the same afternoon at the temperatures indicated in the paper. They were then laid on one side, and the various tests were carried out over a period of approximately one month, starting with those tempered at 150° C. All the curves seemed to be of a similar shape, as would be seen from the diagrams, and, therefore, he did not see how ageing could affect the results to any marked degree.

The methods adopted for finding the modulus of elasticity were somewhat crude, an ordinary wire extensometer being used, but the authors were more interested in the modulus of elasticity for the materials. Had the variation been large, they would have had to modify the stress calculations, but actually, so far as could be seen from the elastic curves—and they were carefully drawn several times—the modulus of elasticity varied so little as to be negligible for purposes of calculation.

The suggestion that a study of the fatigue curves should be made near to one of the critical points was interesting—in fact, the authors had it already under consideration. It would certainly help to " clinch " the point, and perhaps might give some indication of the reason for the phenomena which had been observed.

In further reply, the AUTHORS wished to thank all those who had contributed to the discussion of the paper for the interesting points raised and suggestions offered.

Some of Mr. Sutton's points had already been answered, but with regard to the omission of S/N curves and figures, as in a reply made previously to a similar question, the authors felt that to include this mass of data would use up far too much valuable space in the *Journal*. However, these figures were available for anyone who might be interested.

Professor Thompson asked questions as to the accuracy and reproducibility of the results, and in the verbal reply it was stated that the limits lay within $\pm 0{\cdot}5$ tons per sq. in. To give written confirmation of this, three further sets of fatigue tests had been made since the discussion of the paper, on the following wire specimens :

	Carbon. %	Reduction. %	Tempered at—
(1)	0·55	80	200° C.
(2)	0·79	80	200° C.
(3)	0·79	83	200° C.

The following fatigue limits were obtained, and were contrasted with the originals given in the paper :

	Limiting Fatigue Stress.	
	Recent Determination.	Original Determination.
(1)	± 34·1 tons per sq. in.	± 34·0 tons per sq. in.
(2)	± 29·7 „ „	± 29·5 „ „
(3)	± 39·8 „ „	± 40·0 „ „

These figures should be self-explanatory. At the same time, they seemed to indicate that ageing effects were non-existent, although it must be stated that tests (2) and (3) were made about eight months, and test (1) about five months after the originals, and that it was usually accepted that the greatest changes due to ageing took place within the first few days after treatment.

In reply to Mr. Atkins, the authors wished to point out that the decarburisation was removed from the patented rods before drawing into wire, and not from the wire after drawing ; the wires were tested in the as-drawn condition. The authors were at the present time carrying out parallel tests on ordinary commercial—*i.e.*, decarburised—materials, and it was hoped that the results of these investigations would be published later.

The authors agreed with Dr. Swinden that at present the data obtained were only directly applicable to springs, and that for other purposes many other factors could influence the life and behaviour of a wire in service. His suggestion of investigation of the effect of surface imperfections or notch sensitivity upon tempered wires of varying carbon content and percentage reduction was very interesting, although the authors were of the opinion that it would be a somewhat difficult problem to obtain reliable data. Two lines of attack seemed to be possible :—

(1) Polished wire, *i.e.*, all surface marks removed, compared with the same material carrying artificially made scratches or grooves of known and controlled dimensions, and
(2) Wire in the as-drawn condition compared with the same wire when polished.

Both methods were open to some objection. Artificially made grooves and scratches did not always modify the imposed stresses in the same way or to the same degree as ordinary die marks, and it was almost impossible to compute the size and shape of the normally occurring surface imperfections. However, the second suggested method would give comparison with the worst possible conditions.

Some results which the authors had already published[1] gave an indication of the effect of surface imperfections on wire which had not been tempered. The fatigue limit of a decarburised

[1] *Journal of the Iron and Steel Institute*, 1934, No. II. p. 293.

wire of 0·79 per cent. carbon content drawn to 75 per cent. reduction, could be increased by 25 per cent. when all the surface imperfections were removed by polishing. A similar wire drawn from decarburised rod for 72·5 per cent. reduction gave an increase on polishing of approximately 24 per cent. In view of this question, the authors had carried out some tests on the 0·86 per cent. carbon material drawn to 78 per cent. reduction and tempered at 200° C., which gave the highest observed fatigue limit, ± 45·0 tons per sq. in., in the series described in the present paper. After polishing, the following fatigue results were obtained, but, unfortunately, there was not sufficient material at hand to complete the fatigue curve :

Range of Stress. Tons per sq. in.	Stress Cycles.	Range of Stress. Tons per sq. in.	Stress Cycles.	
± 58·8	122,000	± 53·0	12,370,000	unbroken
± 57·6	108,000	± 50·0	10,245,000	,,
± 56·0	293,000	± 48·0	8,362,000	,,

It would be seen that the fatigue limit lay somewhere between ± 53·0 and ± 56·0 tons per sq. in., and probably was about ± 54·5 tons per sq. in. That represented an increase of about 21 per cent. due to polishing. Hence it would seem that the effect of surface imperfections were not quite independent of the treatment which had been given. That inference, of course, was based on only three results, and, no doubt, further investigation would bring other interesting points to light.

Referring to Mr. Winterbottom's remarks, it was generally accepted that no visible changes could be observed in the microstructure of cold-drawn wires after low-temperature tempering, but that did not preclude atomic changes in the unit crystals or fibres. Correlation of the results of X-ray examination with fatigue results would probably yield valuable information, but, unfortunately, the authors had no X-ray apparatus at their disposal. In view of his suggestions, it was interesting to note that the hoped-for 24 per cent. improvement on polishing had not been quite attained, being computed at about 21 per cent. (see reply to Dr. Swinden), so probably there was some other factor which must be considered, as he suggested.

With regard to his questions about the types of fracture obtained, in all cases the fatigue fractures were flat or the initial crack was flat. In some cases, especially with the wires of higher carbon content, higher percentage reductions and higher temperatures of tempering, after the fatigue crack had started the material tore, giving the appearance of a helicoidal fracture. The torsion fractures were variable, but, generally speaking, the helicoidal fractures were obtained with the material which had received

the higher percentage reductions and which had been tempered at the higher temperatures. As the carbon content was increased helicoidal fractures became more general, and with the highest carbon content they occurred under practically all conditions of treatment.

In reply to Professor Lea, the authors said that they quite agreed that a great deal more work would have to be done before the problems connected with the low-temperature heat treatment of drawn wires could be clarified ; for example, the results which the authors had given in the present paper did not touch on the question of the effect of time of tempering, which might affect the results considerably. The authors had already hinted that there was some "size" effect, that is, the wires of larger diameter seemed to behave differently, in degree, from the wires of smaller diameter. However, they could not agree with Professor Lea that it was desirable to use temperatures of the order of 400° C. for the general tempering of cold-drawn steels. Specifications relevant to the question gave the upper temperature limit as being 350° C. or 375° C., and for general purposes 200° C. to 250° C. was to be preferred. That, of course, did not preclude seemingly extraordinary results being obtained at higher temperatures, such as the 100 per cent. increase in the fatigue properties of a cold-drawn wire by tempering at 400° C. Cold-drawn wire was peculiar material, and the unexpected often occurred. Unfortunately, it was usually impossible to account for such results, and often very difficult to repeat them. If this could be done the findings would probably lead to very important improvements.

The torsional fatigue results which Professor Lea had given for a sample of D.T.D.5A. wire were very interesting, and they seemed to fit in very well with the general results which the authors had given in the paper. From the figures given it would seem that the wire had been drawn to 81 to 82 per cent. reduction. Inspection of Figs. 3 and 5 in the paper (which were for material of similar carbon content) showed that at about that percentage reduction the fatigue curves for the material after tempering crossed those for the material in the as-drawn condition, so that at that point little or no change might be expected in the fatigue limit after heat treatment. That seemed to be confirmed by Professor Lea's results, although the types of stress used were different.

INVESTIGATION OF THE BEHAVIOUR OF METALS UNDER DEFORMATION AT HIGH TEMPERATURES.

PART I.—STRUCTURAL CHANGES IN MILD STEEL AND COMMERCIAL IRONS DURING CREEP.*

By C. H. M. JENKINS, D.Sc., and G. A. MELLOR, M.Sc.
(National Physical Laboratory.)

Summary.

The present work contains the results of a study of the effects of deformation at high temperatures of several varieties of iron and mild steel and shows marked differences in behaviour which are probably related to the mode of manufacture.

The study has been made by microscopical examination of material subjected to creep and short-time tensile tests at high temperatures, the test-pieces being kept free from attack by air by stressing *in vacuo* in special testing units.

The nature of the creep movement has been studied in relation to the structural changes observed both on the surface of a polished test piece and on internal sections. The investigation shows that slip, grain boundary movement, and recrystallisation are prominent features in the mode of deformation at high temperatures. Spheroidisation is also of importance in mild steels below the critical range. Intercrystalline cracking is encountered principally in tests of Armco ingot iron.

Recrystallisation in creep tests proceeds by a continuous action accompanying the deformation and is most prominent in places of maximum deformation.

Section 1.—Introduction.

The investigation described in the present paper has been undertaken in the Metallurgy Department of the National Physical Laboratory as a portion of the programme of the Committee on the Behaviour of Materials at High Temperatures. This work has been carried out under the supervision of the Metallurgy Research Board of the Department of Scientific and Industrial Research. The authors are greatly indebted to these bodies for permission to publish the results of this research. The object of this work is to correlate the structure and constitution of alloys with their behaviour at high temperatures.

* Communication from the National Physical Laboratory, received July 22, 1935.

The present paper describes the results of an investigation into the behaviour of mild steels and commercial irons under the prolonged application of stress at high temperatures. The changes which accompany deformation have been studied by subsequent microscopical examination and are of two types : The first, which proceeds in a material as a result of prolonged exposure to elevated temperatures without deformation, and the second, which is set up by the plastic flow of the material.

The tests described in the present paper have been made *in vacuo* in order to eliminate the oxidising effect of the atmosphere. Even at high temperatures this treatment has not brought about etch-pitting or any appreciable loss of metal from the surface of the test-piece. Thus, after periods of testing lasting for as long as several months the surfaces of highly polished test specimens have been adequately preserved.

Although the primary object of this paper has been the study of the mode of deformation of certain forms of iron and steel at high temperatures, a systematic survey of the strength of these materials over a considerable range of temperature has also been made (300°–950° C.). In addition to creep tests, a series of short-time tensile test specimens have been pulled in an Amsler testing machine adapted for carrying out tests *in vacuo*. The rate of extension of a few of the creep test specimens has not been appreciably less than that of the short-time tensile tests. As far as can be seen the manner of pulling has not resulted in any appreciable difference in the structural behaviour apart from the effects directly ascribable to the duration of testing and the degree of deformation.

Many of the phenomena observed in connection with iron and steel are, of course, similar to those observable in other metals and alloys subjected to stresses sufficient to cause flow at appropriately chosen temperatures, but the change in the lattice structure of iron at a temperature slightly higher than 900° C.—the A_3 point—has a marked effect on the resistance of iron and its alloys to creep. This paper does not deal with the deformation of cold worked material, nor does it deal with the effect of creep on hardened and tempered mild steel.

Section 2.—Experimental Materials and Methods.

Materials.—The present investigation has been made on the following materials : Two mild steels marked *NGK* and *M*, two Swedish (Lancashire) irons marked *S* and *L*, two Armco ingot irons marked *A* and *H* and a sample of carbonyl iron marked *C*. The materials were in the form of hot-rolled or normalised round bars, with the exception of ingot iron *H* and carbonyl iron *C*, which were rolled plates. The chemical compositions of these

materials are given in Table I. and the results of Brinell hardness and short-time tensile tests at room temperature are recorded in Table II.

TABLE I.—*Analyses of Materials.*

Element.	Mild Steel.		Swedish Iron (Lancashire).		Armco Ingot Iron.		Carbonyl Iron.
	NGK. Basic O.-H. Process.	*M.* Acid O.-H. Process.	*S.*	*L.*	*A.*	*H.*	*C.*
Chemical Composition.							
Carbon. %	0·24	0·17	0·017	0·034	0·016	0·022	0·015
Silicon. %	0·05	0·13	0·05	0·024	0·006	0·002	< 0·002
Sulphur. %	0·032	0·018	0·003	0·003	0·017	0·035	< 0·0005
Phosphorus. %	0·016	0·025	0·036	0·056	0·003	0·004	< 0·0008
Manganese. %	0·63	0·685	0·005	Trace	0·030	0·035	< 0·005
Nickel. %	Trace	0·061	Nil	0·004	0·03	0·048	0·011
Chromium. %	Nil	Nil	Nil	0·004	Nil	Trace	Nil
Copper. %	0·09	0·044	Nil	0·005	0·07	0·084	< 0·002
Aluminium. %	Trace	Trace	Trace	Trace	Trace	Trace	Nil
Oxygen by vacuum fusion (carbon reduction). %	0·022 0·022 ..	0·0049 0·0052 ..	0·37 0·18 0·49	0·21 0·43 ..	0·20 0·15 ..	0·23 0·15 ..	0·0020 0·0018 ..
Nitrogen. %	0·008	0·010	0·006	0·004	0·003	0·006	0·003
Spectrographic Analysis.							
Copper	× × ×	× × ×	×	×	× × ×	× × ×	×
Nickel	× ×	× ×	×	–	× ×	× ×	×
Carbon	× ×	× ×	× ×	× ×	× ×	× ×	× ×
Chromium	×	×	–	–	×	×	–
Manganese	× × × ×	× × × ×	×	×	× ×	× ×	×
Vanadium	–	–	–	–	–	–	–
Calcium	× ×	× ×	× ×	× × ×	× × ×	× ×	× ×
Cadmium	Doubtful						
Aluminium	× ×	×	×	×	× ×	×	× ×
Magnesium	× ×	× ×	× ×	× × ×	×	× ×	× ×
Phosphorus	–	–	–	–	×	–	–
Silicon	× ×	× × ×	×	×	–	–	×
Sodium	× ×	× ×	× ×	× ×	× ×	× ×	–
Barium	×	×	× ×	×	× ×	×	–
Molybdenum	Doubtful						
Titanium	Doubtful						

– Indicates element probably absent.
× ,, very faint trace.
× × ,, faint trace.
× × × ,, heavier trace.
× × × × ,, >0·6 per cent. of manganese present (by comparison with chemical analysis).

The mild steel marked *M* in the present series is identical with a 0·17 per cent. carbon steel, the properties of which were previously described in *Engineering Research Board Special Report No.* 1, published by H.M. Stationery Office. In that report the results of tensile, torsion and notched bar tests are given for the temperature range 15°–700° C., of creep tests for the range 250°–500° C., and of fatigue tests for the range 15°–600° C.

No appreciable degree of segregation of the sulphur and phosphorus contents was revealed by the macroscopic examination of longitudinal and transverse sections of round bars or of transverse sections through the plate material. (*See* Figs. 17 to 27, Plates XIX. and XX.). Ingot iron *H* is, however, an exception, showing a marked segregation in the centre of the plate. (*See* Fig. 25.) The position of the segregate is such that test-pieces

TABLE II.—*Mechanical Properties of Materials at Room Temperature.*

Material.	Letter.	Condition.	Brinell Hardness Number. $L/D^2 = 10$.	Yield Point. Tons per sq. in.	Ultimate Load. Tons per sq. in.	Elongation. † %	Reduction of Area. %
Mild steel.	*NGK*	Normalised from 950° C. at N.P.L. ¾ in. round bar.	107	18·0 18·6	25·4 26·4	50 51	67 68
	M	Normalised by makers. 1 in. round bar.	110 127*	16·8*	28·7 28·5*	53 43*	68 68*
Swedish iron.	*S*	As received. 1 in. round bar.	79	14·8 17·8	19·7 20·7	59 51	79 83
	L	As received. ¾ in. round bar.	75	17·3 16·1	20·4 20·4	53 55	77 80
Armco ingot iron.	*A*	As received. 1 in. round bar.	83	14·6 16·1	20·2 19·9	53 63	76 82
	H	As received. ⅜ in. plate.	95	10·2 (*a*) 16·9 (*a*) 13·5 (*b*) 16·4 (*b*)	18·6 18·9 21·2 22·5	46 56 48 40	69 78 71 74
Carbonyl iron.	*C*	As received. ⅛ in. plate.	80	16·4 12·1	17·2 17·1	19‡ 22‡	96 96

* Properties recorded by H. J. Tapsell and W. J. Clenshaw in *Engineering Research Board*, 1927, *Special Report No.* 1.
(*a*) In the direction of final rolling.
(*b*) At right angles to direction of final rolling.
† Specimens 0·358 in. diam. 1 in. gauge length.
‡ Specimens 0·200 in. diam. 1 in. gauge length. Both specimens showed a very marked yield point and broke outside the gauge length.

prepared from this iron consist wholly of this segregated portion, whereas ingot iron *A* is not appreciably segregated. The sample of rolled carbonyl iron shows remarkable uniformity, as would be expected from the chemical analysis and from the mode of manufacture, which consisted of sintering and rolling.

The microstructures of the materials before testing are reproduced in Figs. 18, 19, 21, 22, 24, 26, and 28. The structures of both *M* and *NGK* are typical of steels of these compositions, although the pearlite in mild steel *M* is somewhat banded. The two Swedish irons contain a normal proportion of non-metallic material, and a quantity of finely divided non-metallic inclusions is present in both samples of Armco ingot iron. The grain size of the two Swedish irons is appreciably larger than that of the other materials.

Preparation and Examination of Test Specimens.—The test specimens, which are more fully described in a section on the creep testing apparatus, were polished, after preliminary fine machining, on the whole length of their parallel portion. Before testing, all specimens were etched in a 4 per cent. solution of nitric acid in alcohol, and their microstructures compared with the normal microstructure of the material. After testing the elongation was measured by the extension of a marked length on the parallel portion of the test-piece. In the form of testing apparatus adopted it has not been possible to record continuously the extension of the specimen on a measured gauge length, but observations on the screwed head above the upper tension bar have given an indication of the degree and rate of deformation of the test-piece as a whole. On removal from the creep testing apparatus, the surface of the specimen was neither re-polished nor re-etched for microscopical examination, and the changes observed were the results of alterations of level on the surface of the specimens superimposed on the microstructure revealed by the original etching solution. During the examination of the surface of the test specimen special notice was taken of the degree of slip and the condition of the grain boundaries after deformation, and a careful study of the nature of the fracture was made. The fracture was preserved by a thick electro-deposit of iron applied to the test specimen before sectioning longitudinally for microscopical examination.

Reference to previous researches on the crystalline state of metals shows that rupture may be either definitely intercrystalline or definitely transcrystalline, but the evidence now put forward does not justify any hard and fast classification, except, perhaps, in the case of Armco ingot iron. In addition to a detailed study of the fracture and the surface of the specimen, the condition of the metal in the interior of the test-piece was examined.

The majority of the test specimens have been strained until rupture occurred. The degree of distortion and recrystallisation of the grains, the course of the cracks in relation to the grain boundaries, the presence of cavities and the spheroidisation of the pearlite, where the last named was present, have all received consideration. In all cases the major dimension of the photo-

graphs corresponds to the direction of application of the load (the major dimension is across the page on Plates XIX to XXXVIII.).

The appearance of the polished surface of a broken specimen is not always similar to that observed on longitudinal sections cut axially through the test-piece. For example, the surface may show the presence of slip planes, a widening of the grain boundaries and a general crumpling and corrugation of the surface, but the examination of the interior structure does not necessarily show appreciable change. The surface differences, which are more fully discussed later in the paper, are produced by localised stress set up in partly unsupported crystal grains.

The Creep Test Apparatus.—The apparatus consists of three testing units mounted in a steel frame. Each specimen is loaded

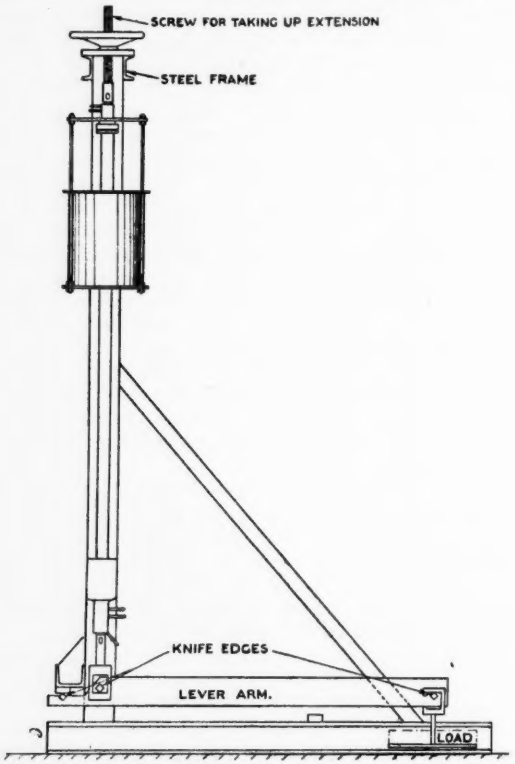

FIG. 1.—Diagram of Creep Testing Unit in Frame.

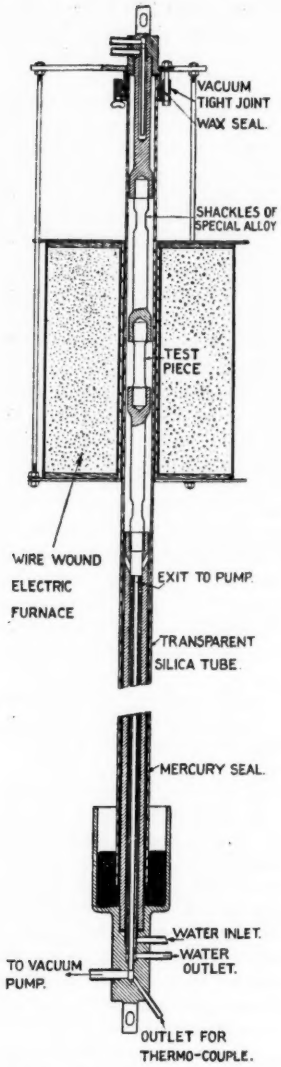

FIG. 2.—Diagram of Interior of Creep Testing Unit.

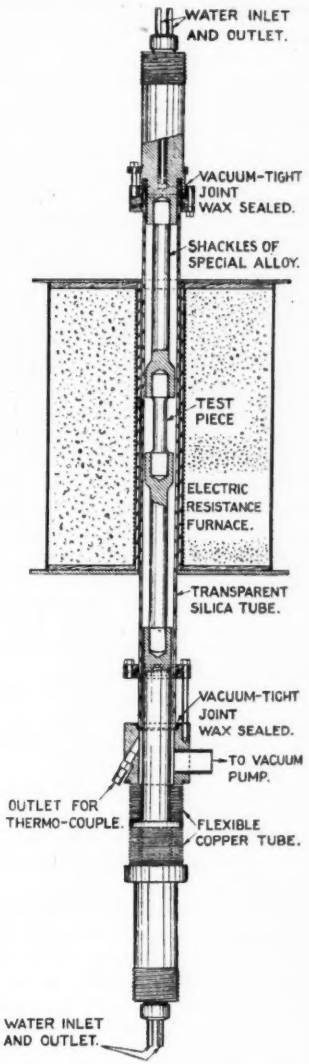

FIG. 3.—Diagram of Vacuum Tensile Testing Unit.

by means of a lever arm having a ratio of either 10 : 1 or 5 : 1. The arrangements for stressing the specimen are shown in Fig. 1 and are similar to those employed by the late Mr. J. H. S. Dickenson[1] in his researches on the creep of metals.

The test specimens, which are of various diameters between 0·20 in. and 0·564 in. according to the size of the original material or to the stress to be applied, are screwed into suitable adaptors to fit the tension bars. The shackles projecting into the furnace are made of a heat-resistant alloy, and these screw into two water-cooled mild steel tubes of heavy gauge, as shown in Fig. 2.

On account of the thinness of the plate of carbonyl iron the minimum load which can be applied to a specimen by this arrangement is too great for tests at the highest temperatures. By the introduction of a counterbalancing system of weights, the load on the test-piece can be reduced to as little as a few pounds without appreciably diminishing the accuracy of loading.

To allow for the extension of the test-piece, an automatic arrangement raises the upper tension bar so that the lever is maintained in a horizontal position, and thus continuous operation of the testing units is possible without manual attention. This reduction in movement of the lever makes it possible for the clearance between the lever and the stops, mounted on the frame, to be reduced to a minimum. On completion of a test by rupture of a specimen the fall of the lower tension bar and lever is not sufficient to cause breakage of the silica tubes. In dropping, the lower tension bar operates a contact which switches off the furnace current. The moment of failure is indicated by the fall in the temperature which immediately occurs, and is recorded by a Cambridge thread recorder.

The three units are individually evacuated by means of two-stage mercury vapour pumps backed by a large rotary pump. The pressure has not normally been in excess of 0·0001–0·0002 mm. of mercury.

The specimen for test is surrounded by a silica tube which is supported at the top of the unit by a vacuum-tight waxed joint. The lower end of this tube dips into mercury and forms another vacuum seal which gives freedom of movement. The general arrangement is shown in Fig. 2, and this diagram indicates that the tube and the lower tension bar are of such lengths that on evacuation there is sufficient height for the mercury to form a barometer column. Allowance is made for the change in the effective load on the specimen resulting from the pressure difference between the evacuated space and the atmosphere.

The test specimen is heated by a concentric tube furnace, controlled at a constant temperature by means of a thermostat, and after the specimen has attained the testing temperature the load is applied. The hot junction of the thermocouple used

to measure the temperature is shielded from radiation by a strip of nickel sheet, and the thermocouple wires are threaded through the central pipe within the lower tension bar. The temperatures recorded on the specimens were checked by readings taken by thermocouples inserted in holes drilled in the parallel portion and threaded ends of a specially arranged test-piece inserted in the creep unit.

Vacuum Unit for Amsler Testing Machine.—The design of the unit for carrying out short-time tensile tests *in vacuo* is shown in Fig. 3 and is arranged for assembly between the jaws of an Amsler testing machine of 30 tons capacity. The lower vacuum-tight joint to the unit consists, in this case, of a flexible copper tube, and not of a barometer column as employed in the creep testing unit. Corrugated copper tubing about 2 in. in diam. is used to form this joint and before commencement of a test the tubing is fully compressed so that sufficient length is obtainable for subsequent extension under load. The load required to extend the tubing is small and corresponds in magnitude to other possible errors, but the effect of evacuation is appreciable as it causes a reduction in the effective tensile load of about 60 lb., for which due allowance is made.

General.—In the following sections a description is given of the general behaviour of mild steel and the various forms of iron under deformation. At the conclusion of the paper the changes in the properties of the materials and the microstructure are discussed in relation to their importance and the mode of deformation. It is of general interest at this stage to note the following values of the load which after continuous application for 10 days at 450° C. will break specimens of these materials.

Mild steel	*M*	16–17 tons per sq. in.	
„ „	*NGK*	13–14	„ „
Swedish iron	*S*	6–7	„ „
Armco iron	*A*	6–7	„ „
Carbonyl iron	*C*	5	„ „

It should be noted from the above results that an almost carbonless iron is inferior in creep resistance to steel containing a small proportion of carbon, and although part of the strengthening is probably due to carbon, the steel contains other hardening elements. It is likely, as observed from other work in hand, that steels of higher carbon content are not superior in creep resistance to the mild steel mentioned above. This fact is of importance, for at room temperature an increase of the carbon content results in increased strength.

The data given above show that steel *NGK* is weaker than steel *M*, although it contains 0·24 per cent. of carbon compared with

0·17 per cent. for the latter material, but the analyses show that the silicon and nickel contents are higher in the stronger material.

The mode of deformation of *M* and *NGK* is found to be very similar, and a similarity also exists between the two samples of Armco ingot iron and between the two Swedish irons and carbonyl iron.

Section 3.—The Behaviour of Mild Steel.

The behaviour of mild steel and other metals under plastic deformation at high temperatures has been studied by Rosenhain and Humfrey[2] and by Rosenhain and Ewen.[3] These authors suggest that failure at high temperatures occurs by intercrystalline cracking produced by the diminished strength of the highly disorientated grain boundary material.

Previous experimental work[4] shows that in mild steel there is an increase in tensile strength to a maximum value at approximately 250° C. Below this temperature the value of a load which results in very small rates of strain approximates to that determined in an ordinary short-time tensile test. In such a test there is insufficient time for appreciable amounts of contemporaneous ageing to be imparted to the specimen. An ageing effect of this type would increase the hardness of the material beyond that produced by the action of "cold work" from straining which is always developed during the course of any ordinary cold tensile test.

Above this temperature the application of a load that is smaller, and at high temperatures very considerably so, than that required to bring about rupture of a short-time tensile specimen may ultimately bring about failure on more prolonged loading.

Mild Steel NGK.—The principal properties of this material as shown by tensile and hardness tests are recorded in Tables I., II. and III., and data from these tables are given in the form of curves in Figs. 4, 5 and 6. Fig. 4 shows the relationship between temperature and load, and against every plotted point the life in days of the particular specimen is indicated. The fully lined curve indicates the strength of the material as disclosed by short-time tensile tests, and the broken line indicates the load-temperature relationship which will bring about rupture after the application of stress for 10 days at a particular temperature. This period of time has been selected so that the materials can be compared conveniently without an undue number of tests exceeding 20 days. Figs. 5 and 6 show for constant temperatures the relation between the load and the expected life of the specimen plotted logarithmically. The points lie on slightly curved lines.

A consideration of Table III. shows that while the strength

under prolonged load slowly decreases with increase of temperature, many of the high-temperature tests show very appreciable elongation, frequently exceeding 50 per cent. At 450° C. and at 850° C., the elongation results are found to be low. At the former temperature the poor ductility may be associated with incipient spheroidisation and at the latter with the proximity

TABLE III.—*Tests on Mild Steel* NGK in Vacuo.

Temperature ° C.	Specimen No. *NGK*—	Machine Used.	Creep Load, or Ultimate Load on Amsler Machine. Tons per sq. in.	Elongation. %	Reduction of Area. %	Duration of Test. Days.	Hr.	Micrograph. See Fig.—
15	271*	Amsler.	25·4	50	67	..	..	29 and 30
350	239	Amsler.	27·7	37	62	..	..	..
400	225	Creep.	19·8	36	62	17	21	..
450	240	Amsler.	21·3	48	73	..	..	31
	273	Creep.	18·0	11†	71	..	2	..
	236	Creep.	15·8	21†	67	2	1	32
	280	Creep.	13·25	31†	65	12	22	33
	272	Creep.	12·0	8†	58	21	12	34, 35 and 36
	312	Creep.	10·0	10†	65	49	0	..
500	268	Creep.	10·0	28†	76	2	18	37
	223	Creep.	9·9	31	75	4	14	38
	238	Creep.	9·9	38	73	7	7	39
550	235	Amsler	13·7	67	81	..	..	40 and 41
	224	Creep	5·9	30†	82	4	22	..
	281	Creep.	4·0	75	72	24	0	42 and 43
650	237	Amsler.	8·24	93	86	..	..	44
	226	Creep.	1·94	69	92	6	0	45
750	242	Amsler.	4·64	75	96	..	..	46
	241	Creep.	0·95	111	92	4	4	..
850	243	Amsler.	4·67	65	93	..	..	..
	246	Creep.	0·50	10	40	31	12	47 and 48
	245	Creep.	0·30	9	Not broken.	81	0	..
950	244	Amsler.	3·35	107	99	..	..	..
	269	Creep.	0·73	54	46	3	17	..
	270	Creep.	0·50	33	29	12	4	49

* Tested in air.
† Broke outside gauge length.
Tests conducted in Amsler machine to give a life of about 2–3 min.

of the upper critical point in steel. It is of interest that the $\alpha \rightarrow \gamma$ change has increased the short-time tensile strength at 850° C. to 4·67 tons per sq. in., whereas extrapolation from lower temperatures would suggest a value of about 3·0–3·5 tons per sq. in. The second mild steel *M* referred to later in the paper does not show this absence of full ductility at 450° C., and it was not examined, owing to the exhaustion of the stock of material, at temperatures as high as 850° C.

The fractures of the test specimens were found to conform to the well-known cup and cone type in all cases when the testing temperature did not exceed 650° C., and in short-time tensile tests even to 850° C. Most of the deformation occurred close to the fracture. At higher temperatures, and especially at the slower rates of creep, deformation occurred as a general tapering

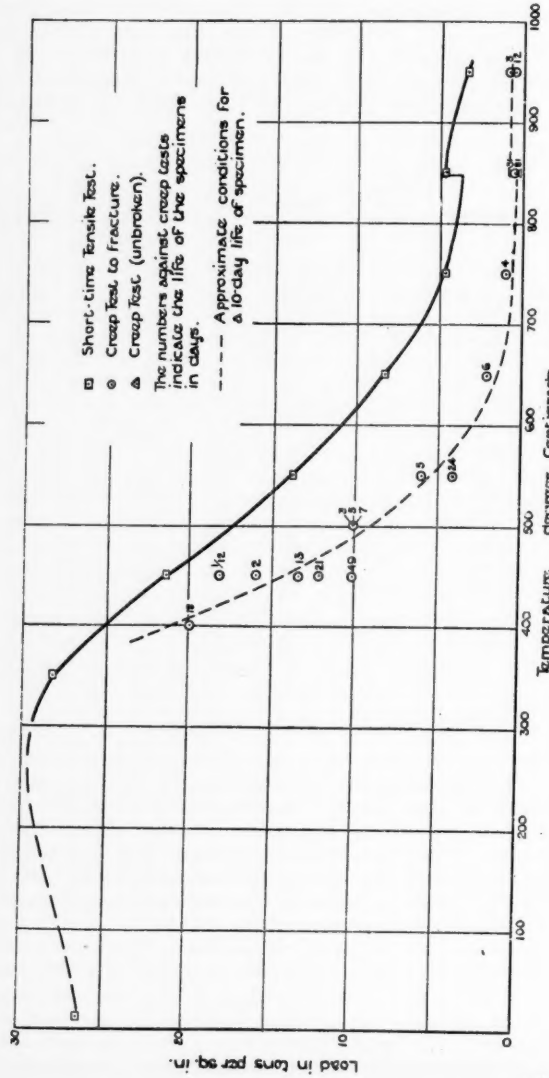

FIG. 4.—Load Temperature Relationship. Mild steel *NGK*.

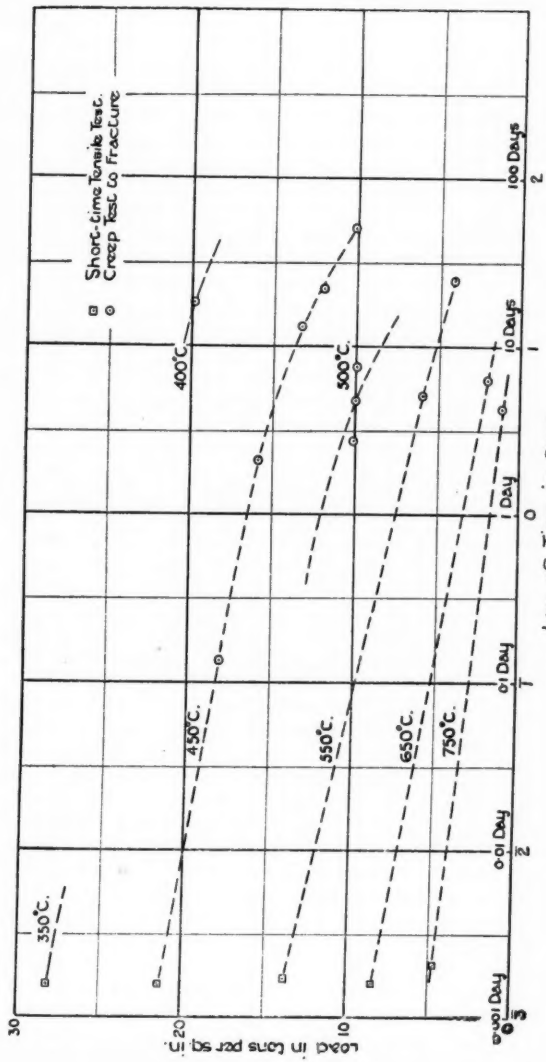

FIG. 5 (continued on Fig. 6).—Load-Life Relationship. Mild steel *NGK*.

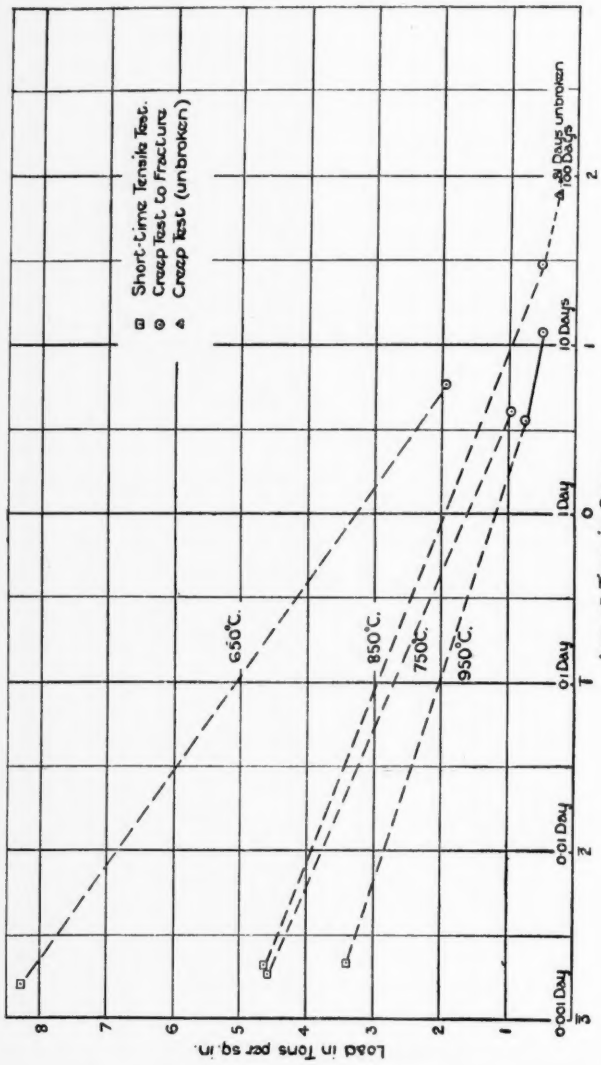

FIG. 6 (continued from Fig. 5).—Load-Life Relationship. Mild steel *NGK*.

of the parallel portion of the test-piece, and the fractures were irregular.

A consideration of the behaviour of the steel as the result of microscopic study indicates that very marked differences occur in the temperature range investigated. Before test the microstructure of this steel, *NGK*, consists of normally distributed pearlite in a matrix of ferrite as is shown in Fig. 18 (Plate XIX.), and straining at room temperature in the course of a short-time tensile test distorts the pre-existing structure as seen either on the surface of the test-piece or on an etched section through the interior. Fig. 29 (Plate XXI.) shows the appearance of the surface after straining and indicates the greatly altered appearance of the ferrite and pearlite. A considerable amount of slipping has taken place in the ferrite, but the masses of pearlite have not greatly changed in shape, although the ferrite grains surrounding them are highly contorted. Fig. 30 shows that the structure near the fracture has been altered still more than that shown in Fig. 29, and the pearlitic areas have been elongated by the plastic flow of the accompanying ferrite grains. At the surface of the specimen (see Fig. 29) the pearlitic areas largely retain their lamellar character, although the internal movement has been considerable. It will be seen in Fig. 30 that in addition to the distortion of the ferrite grains, inequalities of strain have been imposed on individual grains. This residual strain has set up " strained areas " within the ferrite, as revealed by etching the polished section.

In a short-time tensile test at 350° C., the material exhibits increased strength compared with a room temperature test, and the structural changes taking place at these two temperatures are very similar. At low temperatures it appears that movement initially occurs by slip in the ferrite and also takes place in the pearlite, but the general extension of the specimen is not accompanied by gradual movement within the crystals, but by movements initially of a block character. The degree of movement appears to vary from one part of a crystal to another, although slip has occurred by block movements more or less throughout all crystals in the initial stages of deformation, *i.e.*, just after the yield point has been passed.

The contorted character of grains under deformation is, however, better seen at higher temperatures than at lower, such as 350° C. The pearlite, although more rigid than the ferrite, undergoes internal movement which possibly does not occur along the boundary surfaces of the lamellæ, because these appear to be less sharply defined, whereas if movement had taken place in a direction parallel to the lamellæ, it would be expected that the lamellar character would be revealed by the etching solution with greater definition. At temperatures above 450° C. it is

probable that the pearlite is not greatly stronger than the ferrite, and there appears to be little mutual interference.

Fig. 31 shows the microstructure of this steel in the region of the fracture where deformation is greatest, after a short-time tensile test at 450° C. The duration of this test at 450° C. was about 2–3 min., but even with this short time of test it was necessary to heat the specimen for about 1–2 hr. above 300° C. to enable the test to be carried out. The pearlite and ferrite are highly contorted, and it is probable that recrystallisation of the ferrite grains has commenced. The presence of identifiable new grains can be seen after tests at 500° C. or, if considerably longer times of loading are given to the specimen, at 450° C. The steel does not show, between room temperature and 400° C., any very marked difference in the mode of deformation. In the absence of deformation at 450° C. no rapid alteration in structure develops, but if deformation is applied recrystallisation of the ferrite is apparent even after 2 days (see Fig. 32). By the application of a lighter load the specimen life is increased to 12 days, and not only is recrystallisation of the ferrite well marked, but there is also appreciable spheroidisation of the pearlite; this is particularly noticeable at the fracture, where deformation is greatest (see Fig. 33). This micrograph is of interest in that it also shows an exceptional case of the occurrence of intercrystalline cracking in a few of the grain boundaries.

The effect of the degree of deformation on the rate of spheroidisation at 450° C. is more clearly shown in a longer time test (21½ days under a load of 12 tons per sq. in.). Figs. 34, 35 and 36 (Plates XXI. and XXII.) show the condition of the pearlite under three conditions, (*a*) close to the fracture where spheroidisation has progressed far, (*b*) ½ in. from the fracture where spheroidisation is commencing, and (*c*) in unstressed material heat-treated with the test-piece, where no appreciable spheroidisation has occurred, although the cementite at the boundary of the pearlite grain is receding from the neighbouring ferrite grains. Globules of cementite have thus been formed. This shrinkage of the pearlite lamellæ is one of the characteristic features of the initial stages of low temperature spheroidisation. In the temperature range 500°–550° C. spheroidisation occurs rapidly under the action of deformation. Recrystallisation of some of the ferrite is also accompanied by grain growth of other parts. This last-named effect is indicated in Fig. 37, where the crinkled and interlocked character of the ferrite grain boundaries after a test lasting only two days is clearly shown. The generation of new grains and the incipient spheroidisation of the pearlite is seen. Fig. 39 also shows the character of the changes which occur at 500° C., the reduced strength of the pearlite relative to the ferrite is apparent and the development of new crystals after

7–8 days is much further advanced than in the structure shown in Fig. 37.

The correlation of the surface appearance with the internal microstructure has been made by an examination of longitudinal sections through test-pieces polished close to the original surface. An example of such a section is shown in Fig. 38. Some of the crystals situated close to the surface have parted one from the other so that an inspection of the surface would suggest that the material was prone to intercrystalline brittleness. This widening of the grain boundaries at the surface is of common occurrence, and suggests that the crystals at the surface do not distort as much as those further removed from the outside of the test-piece. Care is therefore necessary in interpreting the results derived from surface examination.

The difference between the microstructure of the surface of the material near the fracture and that found $\frac{1}{8}$ in. away is again revealed in Figs. 40 and 41 (Plate XXIII.). Such differences were produced as a result of a tensile test at 550° C. lasting only 2–3 min. In this specimen movement by slip is noticeable in the grains at the original surface, whereas a section through the test-piece showed that slip was accompanied by recrystallisation, which forms the visible indication of internal movement in the ferrite. This recrystallisation is probably a continuous process whereby new grains form and again continue to be deformed and re-annealed in their own turn. The new grains thus formed grow rapidly. At 550° C. the breakdown of the pearlite under the action of strain is very rapid, and in this short-time tensile test-piece spheroidisation has possibly commenced, but physical disruption of the lamellæ might also account for the absence of the characteristic appearance of the pearlite grains

The effect of a test lasting 24 days under a load of 4 tons per sq. in. at 550° C. is very pronounced. The microstructure no longer, as at lower temperatures, shows definite pearlite grains, but the carbide is now present as particles of spheroidised cementite interspersed among the recrystallised ferrite. Fig. 42, at high magnification, shows the ferrite grains, the dispersed carbide and the location of cavities. In the unstressed end of the same test-piece (see Fig. 43) the degree of spheroidisation is considerable, but the pearlitic areas have retained their original shape.

At 650° C. a most pronounced change occurs in the microstructure after a short-time tensile test. In Fig. 44 the ferrite has completely recrystallised, and the pearlitic areas, although still identifiable, are considerably drawn out. The presence of cavities and of cracks—principally parallel to the direction of deformation—is also apparent. In a longer time test lasting six days (Fig. 45, Plate XXIV.) the breakdown of the former pearlitic areas has taken place. The appearance of these

specimens (Figs. 44 and 45) suggests that the pearlitic areas are as weak as the ferrite at 650° C., but it should be noticed that the ferrite has recrystallised, although traces of the old grains are discernible.

The effect of the lower change point in steel, which occurs at 690°–720° C. is to bring about a discontinuity in the properties. As is well known, at 750° C. ferrite and austenite form the normal constituents present in the steel instead of ferrite and pearlite at lower testing temperatures. The carbon of the steel is dissolved in the austenite and is retained in that state until the completion of the test, when, on cooling, the austenite present at the lower critical point is changed to pearlite. There is also a very considerable degree of recrystallisation in the ferrite which is not affected by the lower critical point. A further increase of the testing temperature to 850° C. causes this steel to be wholly changed to austenite (*i.e.*, γ-iron containing carbon in solid solution). The change not only brings about complete and uniform solution of the carbon, but also causes complete recrystallisation.

After a short-time tensile test at 750° C., the microstructure, as depicted in Fig. 46, exhibits very marked recrystallisation. The former austenitic areas are darker than the ferrite unchanged at the lower critical point. These areas consist of cementite and finely recrystallised ferrite.

The effect of a slight leakage of air on the appearance of the polished surface is shown in Figs. 47 and 48. The specimen which had been 31½ days at a temperature of 850° C. exhibited marked relief effects which might be related to slip in a few of the crystals and also shows a very marked widening of the grain boundaries. A cross-section (Fig. 48) shows that decarburisation has occurred to a depth of 0·003 in. The pearlite which has separated from the austenite after the conclusion of the test is lamellar in character. This structure would result from the normal cooling of the specimen in the furnace of the creep unit.

Fig. 49 relates to a specimen tested at 950° C. The specimen shows a coarsely separated form of pearlite, but there is no clear evidence that at the testing temperature the original large austenitic grains, from which the constituents have been subsequently derived on cooling, possess any general intercrystalline weakness.

Mild Steel M.—The second sample of mild steel referred to as *M*, has been tested at various temperatures not exceeding 750° C., and the composition and properties of this material are recorded in Tables I., II. and IV. The results of short-time tensile tests and of creep tests are given graphically in Figs. 7 and 8. Some of the results given in Table IV. and in Fig. 7 have already been published.[5] In Fig. 7 a curve is also given setting

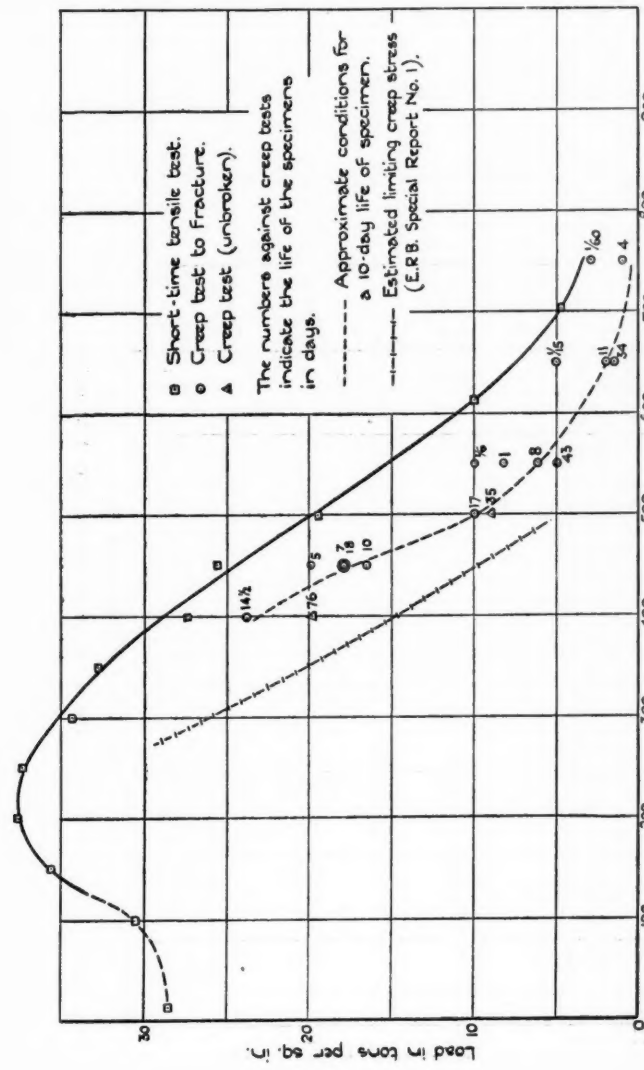

FIG. 7.—Load-Temperature Relationship. Mild steel *M*.

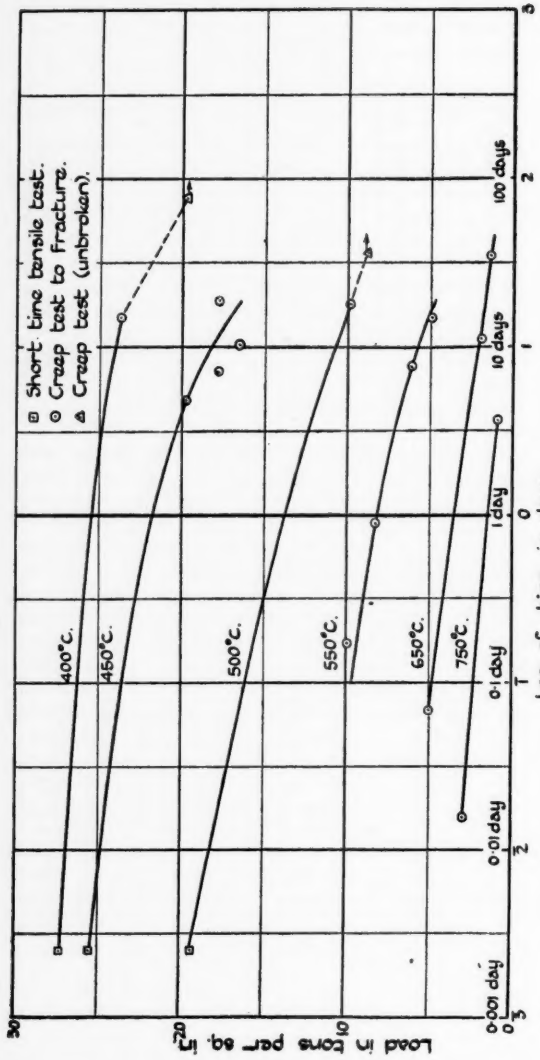

FIG. 8.—Load-Life Relationship. Mild steel *M*.

forth the values of the estimated limiting creep stress as recorded in that report, together with a curve showing the stress/temperature relationship for a 10-day life as deduced from the present results. From a comparison of the latter curve with that shown in Fig. 4, it appears that steel *M* is slightly stronger than steel *NGK* under prolonged loading.

TABLE IV.—*Tests on Mild Steel* M in Vacuo.

Temperature ° C.	Specimen No. *M*—	Machine Used.	Creep Load, or Ultimate Load on Amsler or Buckton Machines. Tons per sq. in.	Elongation. %	Reduction of Area. %	Duration of Test. Days.	Duration of Test. Hr.	Micrograph. See Fig.—
15	111	Amsler.*	28·7	53	68	..	..	..
16	‡	Buckton.*	28·5	43	68	..	..	..
100	‡	Buckton.*	30·5	26	63	..	..	..
150	‡	Buckton.*	35·6	24	52	..	..	..
200	‡	Buckton.*	37·6	24	51	..	..	..
255	‡	Buckton.*	37·3	28	52	..	..	..
300	‡	Buckton.*	34·3	37	59	..	..	..
350	‡	Buckton.*	32·7	37	61	..	..	..
400	‡	Buckton.*	27·3	40	69	..	..	..
	70	Creep.	23·8	33	53	14	12	..
	65	Creep.	19·8	9	Not broken.	76	..	..
450	‡	Buckton.*	25·5	43	71	..	..	..
	63	Creep.	19·8	29	64	4	18	..
	71	Creep	17·8	44	58	18	10	50, 51, 52 and 53
	124	Creep.	17·8	18	58	7	1	54
	100	Creep.	16·5	27	59	10	2	..
500	‡	Buckton.*	19·4	43	73	..	..	..
	74	Creep.	9·9	55	74	17	12	55 and 56
	95	Creep.	8·9	2	Not broken.	34	16	57 and 58
550	98	Creep.	9·9	32†	81	..	4 hr. 10 min.	..
	61	Creep.	8·2	35	78	..	21 hr. 30 min.	59
	62	Creep.	6·12	53	79	7	14	..
	76	Creep.	4·94	45	79	43	6	60
615	‡	Buckton.*	9·9	65	93	..	..	..
650	77	Creep.	5·0	48	90	..	1 hr. 37 min.	61
	114	Creep.	1·94	32	65	10	22	..
	117	Creep.	1·46	38	58	33	18	62, 63 and 64
704	‡	Buckton.*	4·7	83	96	..	..	..
750	75	Creep.	2·9	118	83	..	0 hr. 23 min.	65
	116	Creep.	0·94	82	78	3	15	66

* Tested in air. Rate of loading on Buckton machine, 4 tons per sq. in. per min. The specimens which were tested in the Buckton machine have not been examined microscopically.

‡ H. J. Tapsell and W. J. Clenshaw, *Engineering Research Board*, 1927, *Special Report No.* 1.

† Broke outside gauge length.

Fig. 7 shows the rise in the tensile strength curve to a maximum at 250° C., and the fall in strength which occurs on increase of the

testing temperature. The curve is very similar to that shown for *NGK* in Fig. 4.

At temperatures up to 650° C. the specimens broken in the creep tests show considerable necking, and below 550° C. the fractures are of the cup and cone type. At higher temperatures than these the necking is replaced by a gradual tapering of the test-piece to the point of fracture, and the fractures are coarse and of no definite type.

The microstructure of the material before test is shown in Fig. 19 (Plate XIX.) and the structures of selected specimens after test in Figs. 50 to 66 (Plates XXV. to XXVIII.). In this group the appearances of typical specimens both on the surface and in the interior of the metal are shown.

In a specimen which was pulled at room temperature a considerable amount of distortion of the surface of the crystals has occurred, and some crystals show an appreciable amount of slip.

A creep test lasting for 76 days at 400° C. without rupture of the specimen was discontinued and the specimen was examined after it had extended 9 per cent. A comparison of the microstructure in the stressed portion of the specimen with that in the unstressed end of the test-piece shows that in both cases the ferrite and the pearlite are unchanged.

A test specimen which ruptured under a load of 17·8 tons per sq. in. at 450° C. after 18 days shows that the ferrite is distorted near the fracture, that the pearlite is commencing to spheroidise, and new grains are beginning to form (see Figs. 50, 51 and 52). The unstressed test-piece shows, however, little change in microstructure (Fig. 53), except that the carbide at the edge of the pearlite grains forms a continuous margin around the pearlitic areas after exposure for 18 days at 450° C.

The surface examination of a specimen which broke after seven days at 450° C. shows that there is a little slip on the surface but appreciable widening of the grain boundaries (see Fig. 54). The pearlite is also elongated, but is not clearly defined.

The condition of the surface of a specimen tested for a longer time at 500° C. shows a very marked separation of the grain boundaries at right angles to the direction of pulling (Fig. 55). The material was, however, quite ductile, the extension being 55 per cent. on the normal gauge length. Subsequent microscopic examination of the interior of the specimen revealed no general intercrystalline cracking, although a few cavities had developed, as seen in Fig. 56. This figure also shows that marked recrystallisation of the ferrite and spheroidisation of the pearlite have occurred close to the fracture. In a further and unbroken specimen stressed for 34 days, when 2 per cent. extension had occurred, spheroidisation in both the parallel portion and the

unstressed end of the test-piece had progressed to a similar degree (see Figs. 57 and 58). It is apparent that slight deformation does not appreciably increase the rate of spheroidisation.

The internal structure of a specimen tested at 550° C., which ruptured after 21½ hr. under a load of 8·2 tons per sq. in., reveals very considerable rearrangement of both the ferrite and pearlite grains, as is shown in Fig. 59. The ferrite has also recrystallised in the form of new grains which are related to the main grains after the manner of sub-boundary structures. The pearlite is considerably spheroidised and shows little evidence of its former lamellar character. The unstressed end of another test-piece shows little spheroidisation after a much longer exposure to this temperature (see Fig. 60).

In the creep test of short duration at 650° C., the internal material is found to be almost completely recrystallised, although some of the original boundaries are traceable, as in Fig. 61. The pearlite is also fairly fully spheroidised and no intercrystalline cavities are identifiable. With longer testing periods at this temperature cavities develop, especially in the highly deformed region near the fracture (see Fig. 62), but no evidence of intercrystalline cracking is found. These microstructures indicate that complete recrystallisation under the action of stress has occurred, as the original grain boundaries are not traceable. A comparison of Figs. 63 and 64 shows that the original pearlitic structure has completely broken down under the action of deformation and prolonged exposure at 650° C., and the spheroidised carbide particles are distributed at random throughout the deformed material, which consists of recrystallised ferrite grains. In the unstressed specimen the original form of the pearlite grains is evident, but these pearlitic areas are elongated in a manner prescribed by the original banding, as shown in Fig. 19.

Tests at 750° C., similar to the previous tests on steel *NGK*, were made above the temperature of the lower critical point in mild steel (about 690°–720° C.). Fig. 65 (Plate XXVIII.) shows the condition of the material at the fracture after a rapid test in a creep unit at 750° C. The austenitic areas at the temperature of test have been reconverted into ferrite and pearlite, and the whole of the remaining ferrite has been completely recrystallised under the action of stress.

Fig. 66 shows the interior of a test-piece which broke after 3½ days at 750° C. The presence of lamellar pearlite indicates the breakdown of the austenite on cooling. Complete recrystallisation has occurred and there is no trace of the original structure shown in Fig. 19. Numerous cavities are found between the ferrite and pearlite grains, and were probably situated at the junction of ferrite and austenite grains (at the testing temperature).

Section 4.—The Behaviour of Swedish Iron (*S* and *L*).

The chemical composition and mechanical properties of these two samples of Swedish (Lancashire) iron are recorded in Tables I., II. and V., and some of the results of sample *S* are shown graphic-

Table V.—*Tests on Swedish (Lancashire) Irons* in Vacuo.

Temperature °C.	Specimen No.	Machine Used.	Creep Load, or Ultimate Load on Amsler Machine. Tons per sq. in.	Elongation. %	Reduction of Area. %	Duration of Test. Days.	Hr.	Micrograph. See Fig.—
15	*S* 145*	Amsler.	19·7	59	79	..	..	67 and 68
	L 128*	Amsler.	20·4	42	72	..	..	..
350	*S* 233*	Amsler.	25·7	34	56	..	..	..
	S 73	Creep.	15·8	29	74	27	12	..
400	*S* 107	Creep.	11·8	31†	73	1	3	69 and 70
	S 68	Creep.	9·9	48	83	26	0	..
	L 125	Creep.	9·9	40	79	4	0	71
450	*S* 146	Amsler.	16·2	83	78	..	..	..
	S 72	Creep.	7·9	35	81	2	20	72
	S 92	Creep.	5·8	55‡	82	26	2	73
500	*S* 96	Creep.	3·94	51	86	24	12	..
	L 108	Creep.	3·9	62‡	89	14	6	74
550	*S* 147	Amsler.	10·45	90	88	..	..	75
	S 69	Creep.	3·9	115	86	..	18	76
	S 78	Creep.	2·9	34‡	87	8	2	77
600	*S* 101	Creep.	3·94	57	86	..	0·35	..
	S 99	Creep.	2·94	80	91	..	18	..
650	*S* 228	Amsler.	5·3	132	90	..	..	..
	S 79	Creep.	2·9	58‡	93	..	0·45	78 and 79
	S 102	Creep.	0·94	27	39	28	16	80
	L 127	Creep.	0·94	20	29	28	2	..
750	*S* 229	Amsler.	2·8	127	95	..	..	..
	S 103	Creep.	0·94	80	79	..	5	..
850	*S* 150	Amsler.	1·55	112	97	..	..	81
	S 109	Creep.	0·30	72	89	1	22	82
950	*S* 230	Amsler.	3·33	138	88	..	..	..
	S 231	Creep.	0·50	17	19	8	5	83
	S 123	Creep.	0·29	4	Unbroken	27	0	..
	L 126	Creep.	0·5	22†	13	8	20	..

* Tested in air.
† Broke outside gauge length.
‡ Approximately.

ally in Figs. 9, 10 and 11. A comparison of the two irons shows that at room temperature the sample *L* is slightly stronger than that marked *S*, but in creep it is weaker at 400° and 500° C. At 650° and 950° C. the strengths of the two materials are not appreciably different. The curves plotted in Fig. 9 show the relationship of load, as given by short-time tensile tests, to the temperature of testing, and the connection between load and temperature for an effective life of the test specimen of 10 days. The strength in short-time tensile tests probably attains a maximum value at about 250° C., in accordance with the general behaviour of low carbon steels and iron. The marked increase in strength at the $\alpha \rightarrow \gamma$ change point is apparent from the result of both creep and tensile tests at 850° and 950° C. The load/time

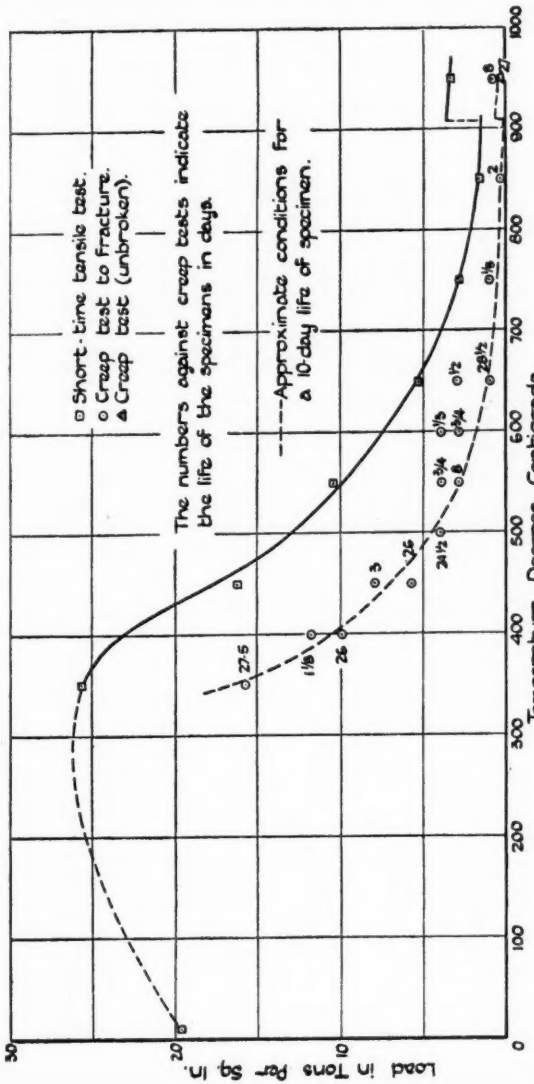
Load in Tons per Sq. In.
Temperature Degrees Centigrade.
Short-time tensile test.
Creep test to fracture.
Creep test (unbroken).
The numbers against creep tests indicate the life of the specimens in days.
Approximate conditions for a 10-day life of specimen.

FIG. 9.—Load-Temperature Relationship. Swedish iron *S*.

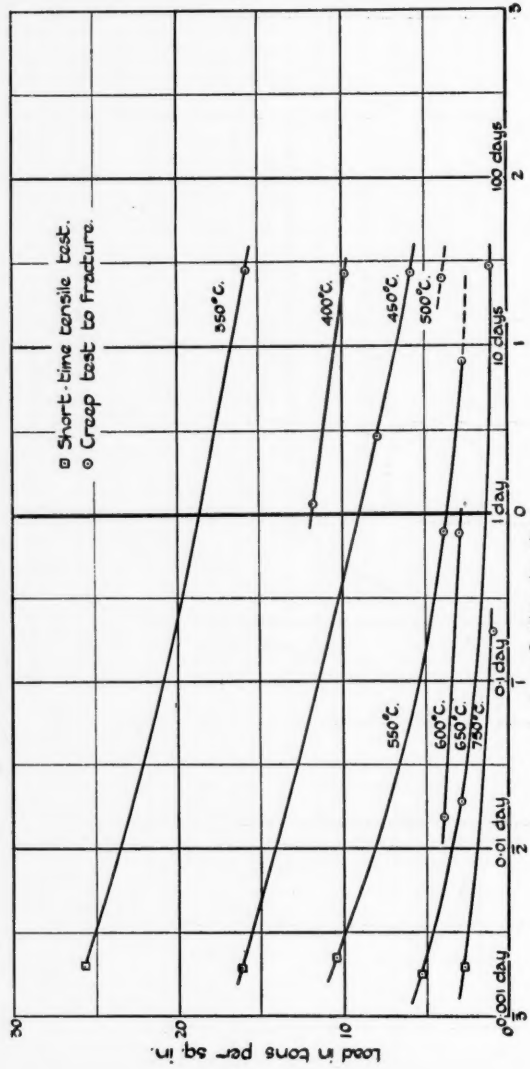

FIG. 10 (continued on Fig. 11).—Load-Life Relationship. Swedish iron *S*.

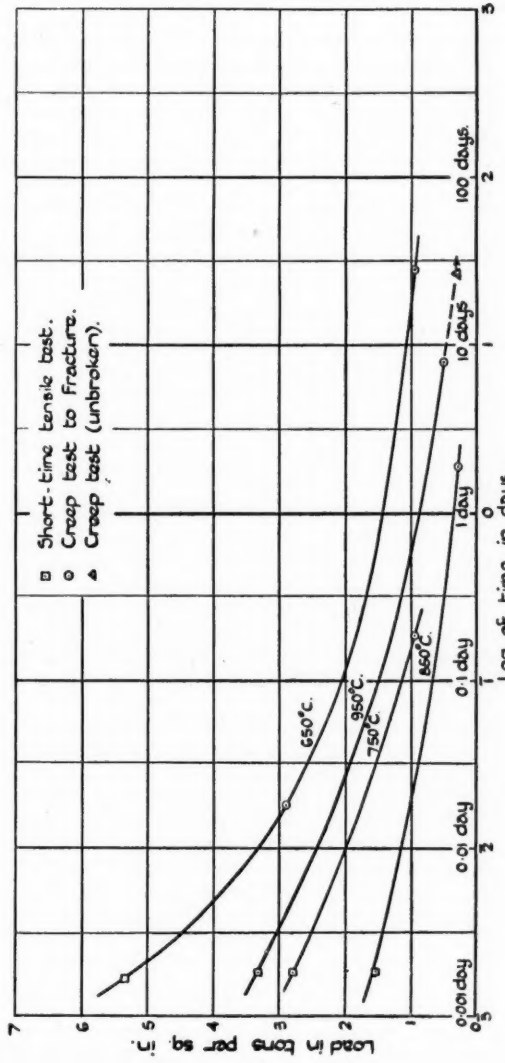

FIG. 11 (continued from Fig. 10).—Load-Life Relationship. Swedish iron 8.

relationships for each testing temperature are given in Figs. 10 and 11.

All test-pieces which have been pulled below 550° C. have fractures of the cup and cone type, although they are less regular in Swedish iron than in the other materials considered. Necking occurs near the fractures at these lower temperatures, but is replaced by a more gradual tapering of the parallel portion as the temperature rises. At higher testing temperatures fractures become irregular.

The microstructures of the materials *S* and *L* before testing are given in Figs. 21 and 22 respectively. The changes in structure as a result of either creep or tensile tests are considered together in the text and micrographs Figs. 67 to 83 (Plates XXVIII. to XXXII.). In contrast with mild steel and with Armco iron, the deformation of Swedish iron and of carbonyl iron is characterised by a marked development of slip, and at higher temperatures general recrystallisation. In Swedish iron, in Armco ingot iron, and in carbonyl iron the material consists of one phase, and only the ferrite, together with the possible interfering effect of non-metallic inclusions and occasional traces of pearlite in the Swedish irons, need be considered. The behaviour of this pearlite has not been studied.

Microscopic examination after testing at room temperature indicates that the planes of slip are very numerous, occurring in some crystals in two or three directions, as is shown in Fig. 67 (material *S*). A longitudinal section through the fracture of the same specimen (Fig. 68) shows marked elongation of the ferrite grains which leads to strongly marked strain-etch patterns. In addition to strain-lines it appears that some of the crystals show strain-bands which differ markedly from the remainder of the crystal.

After a test at 400° C., a polished surface of *S* exhibits sub-division of one or more crystals into smaller units, apparently accompanied by the commencement of recrystallisation (see Fig. 69). A longitudinal section through the same specimen at the fracture shows a very marked elongation of individual grains and of cavities between the grains (Fig. 70).

Fig. 71 shows a longitudinal section through a specimen of material *L* tested at 400° C., and it can be seen that the crystals adjacent to the fracture are highly distorted and have become twisted together as a result of deformation. It is probable that new crystals of ferrite have been formed in the highly strained areas and there is a tendency for some cavities to be drawn out in the direction of the deformation.

After test at 450° C., some crystals on the surface of a specimen of material *S* show well-developed slip in two directions (Fig. 72), but the slip planes appear to be less clearly defined. The grain

boundaries at the surface also show widening. A longitudinal section of a further specimen tested at this temperature until rupture occurred after 26 days showed that the ferrite grains had undergone recrystallisation by a process of subdivision recalling sub-boundaries. The recrystallisation shown in Fig. 73 should be compared with that shown in Fig. 71 ; the comparison indicates that the shorter time and slightly lower temperature of test does not result in the development of much more than strain effects.

Marked recrystallisation is revealed by a specimen of *L* tested at 500° C. (Fig. 74). This micrograph shows the disintegration of several large non-metallic inclusions into a series of disjointed fragments. The resulting space between the particles has become filled by flow of the ferrite. The grain development appears to be little affected by the presence of these inclusions.

At 550° C., the fibrous character of the deformed material close to the fracture is more readily revealed (see Fig. 75). The shape of the deformed crystals is related to the longitudinal tearing which results in the formation of individual fibres, but there is no trace of any cracking at right angles to the direction of pulling. During test this specimen had elongated to the extent of 90 per cent. on the gauge length. It is apparent that the crystals are pulled into straight fibres, whereas at low temperatures (Fig. 70) there is a greater tendency for the crystals to be contorted. The character of the fibrous material is seen in greater detail in Fig. 76. Here the newly-formed grains within the fibres are drawn down in a longitudinal direction, and there are discontinuities between the fibres.

The slip bands on the surface of a specimen tested at 550° C. (Fig. 77) are less clearly defined than at lower temperatures.

The character of the slip formation in the ferrite grains after deformation at 650° C. is shown in Fig. 78. The slip planes are markedly stepped and are not sharply defined, and in one crystal they occur in more than one direction. The appearance of a longitudinal section through the fracture (Fig. 79) of the same specimen is markedly different from that of the surface and shows complete recrystallisation. The new grains which have been formed as a result of the deformation have ceased to be elongated, as owing to the higher testing temperature they have grown after deformation.

The material is divided into fibres which probably correspond to the original crystal grain size, and the formation of these fibres is influenced by the non-metallic inclusions. Numerous cavities have developed within the metal, particularly in the region of the fracture, but there is no transverse intercrystalline cracking such as would indicate a general weakness at high temperatures.

The appearance of the surface of material *S* after a test at 650° C. lasting 28 days is shown in Fig. 80. This photograph, which should be compared with Fig. 78, relates to a long-time test which shows no deformation by slip, whereas the short-time tests at the same temperature have produced slip. In the longer time test the boundaries have been considerably widened at the surface where recrystallisation of the ferrite can also be seen.

TABLE VI.—*Tests on Armco Ingot Irons* in Vacuo.

Temperature °C.	Specimen No.	Machine Used.	Creep Load, or Ultimate Load on Amsler Machine. Tons per sq. in.	Elongation. %	Reduction of Area. %	Duration of Test. Days.	Duration of Test. Hr.	Micrograph. See Fig.—
15	*A* 112*	Amsler.	20·4	44	73	..	..	..
	*H*173*(*a*	Amsler.	18·5	46	69	..	..	84
300	*A* 49	Creep.	15·8	61	71	28	20	..
350	*A* 163	Amsler.	23·5	52	64	..	..	85
	A 45	Creep.	13·8	60	75	1	22	..
	A 41	Creep.	11·8	21	Unbroken	135	0	..
400	*A* 44	Creep.	11·8	61	67	..	6	..
	A 46	Creep.	9·8	40	49	14	22	..
	H 172(*a*)	Creep.	9·9	34	51	..	23	86
450	*A* 164	Amsler.	15·1	61	81	..	..	..
	A 30	Creep.	9·9	58	Not measured	..	2	..
	A 40	Creep.	8·8	38	"	..	12·5	..
	A 58	Creep.	5·9	36	33	41	0	..
500	*A* 80	Creep.	8·9	64	64	..	0·5	..
550	*A* 165	Amsler.	8·6	55	78	..	..	89
	A 34	Creep.	5·9	27	39	..	2·1	..
	A 31	Creep.	3·9	34	Not measured	3	15	..
	H 174(*a*)	Creep.	3·9	17	21	2	16	90
600	*A* 82	Creep.	5·94	40	52	..	0·24	91
650	*A* 166	Amsler.	5·2	69	75	..	..	92 and 93
	A 56	Creep.	3·9	59	50	..	0·15	..
	A 28	Creep.	0·94	2	6	26	12	94
750	*A* 152	Amsler.	3·21	50	54	..	..	..
	A 52	Creep.	0·94	8	13	3	1	..
	A 171	Creep.	0·44	5	5	32	15	95
	H 181(*b*)	Creep.	0·94	2	3	1	2	..
850	*A* 157	Amsler.	2·26	70	66	..	..	96
	A 54	Creep.	0·44	13	18	1	4	..
	A 93	Creep.	0·29	11	20	1	22	97
950	*A* 155	Amsler.	3·17	44	45	..	..	..
	A 55	Creep.	0·44	12	8	3	1	..
	A 90	Creep.	0·24	11	8	14	12	100
	H 175(*a*)	Creep.	0·44	15	9	3	3	99

* Tested in air.
(*a*) Axis of test-piece parallel to direction of final rolling.
(*b*) Axis of test-piece at right angles to direction of final rolling.

The structure of the interior of a short time tensile test-piece is shown in Fig. 81. The recrystallisation at 850° C. is the result of deformation and not of the change point in iron as occurred in the two steels *M* and *NGK*. The unusual and apparently segregated appearance depicted in Fig. 81 is the result of marked grain growth of material recrystallised by the deformation. The surface of another specimen tested at 850° C. does not reveal

slip, but recrystallisation has apparently occurred although the former grain boundaries are still well marked (Fig. 82).

The structure of the interior of a specimen after test at 950° C. is shown in Fig. 83. The material has been transformed by passage through the A_3 change point in iron, and the new γ grains have become relatively large. Numerous cavities and cracks occur in the grain boundaries and some also lie in the grains themselves, but there is no certain information whether their presence is related to a transcrystalline weakness or to the sites of former grain boundaries. No evidence was obtained of general intercrystalline breakdown.

SECTION 5.—THE BEHAVIOUR OF ARMCO INGOT IRON. (*A* AND *H*.)

The chemical composition and mechanical properties of these materials at room temperature are recorded in Tables I. and II., and the results of creep and tensile tests at higher temperatures are given in Table VI. Test-pieces of material *H* cut parallel to the direction of final rolling have proved at room temperature to be weaker than those cut at right angles to this direction, but in this latter direction the material is equal in strength to material *A*. A complete comparison of the relative strengths of materials *A* and *H* at high temperatures has not been made, but in the tests so far carried out no difference in behaviour has been observed. The results of creep and tensile tests of material *A* are shown graphically in Figs. 12, 13 and 14.

The strength of materials *A* and *H* should also be compared with the previously published figures relating to a sample of normalised Armco ingot iron investigated by H. J. Tapsell and W. J. Clenshaw.[5] The principal properties are given below:

Temperature °C.	Estimated Limiting Creep Stress. Tons per sq. in.	Ultimate Stress in Short-Time Tensile Test. Tons per sq. in.	Elongation in Short-Time Tensile Test. %
200	28·0	29·3	26·5
250	21·4	28·9	30
300	16·3	27·9	38·5
350	12·2	25·3	44·5
400	8·4	20·0	51
450	5·0	16·5	51
500	..	12·1	49
600	..	6·8	34
700	..	4·1	53·5

From a comparison of these figures with those given in Fig. 12, it will be seen that in short-time tensile tests up to 700° C., and in creep tests at 300° C., 400° C. and 500° C., the above material is very slightly stronger than material *A*. The value of the estimated limiting creep stress given in the above table is not greatly different from that of the load required to break material *A* in ten days at the same temperature.

The cup and cone type of fracture is found in all creep and short-time specimens tested below 400° C. At this temperature there is a change to a coarse and irregular form of fracture in creep specimens, and at higher temperatures all fractures are of this type. At low testing temperatures specimens show very considerable deformation near the fracture. At temperatures higher than 550° C. the values determined for the elongation and reduction of area are very low except for specimens which have ruptured after tensile tests of a few minutes duration, but even in these cases above 550° C. there is very little necking or tapering of the parallel portion near the fracture.

The microstructures of *A* and *H* in the condition prior to test are depicted in Figs. 24 and 26, and the changes produced by deformation are illustrated by Figs. 84 to 100 (Plates XXXII. to XXXV.). Fig. 84 shows the surface of a specimen of material *H* after a short-time tensile test at room temperature and indicates that slipping occurs in more than one direction in individual crystals. The form of slip in one of the crystals suggests, perhaps, that the ferrite is twinned. The changes in the microstructure of the interior of the specimen are similar to those already described for Swedish iron (See Fig. 68).

A greater development of strain lines is shown by a test carried out at 350° C.; some of these lines are inclined at an angle of 45° to the direction of pulling, as in Fig. 85. This micrograph suggests that there is a large amount of included non-metallic material which does not influence the general character of the deformation.

A longitudinal section through a specimen tested at 400° C. (Fig. 86), shows that the mode of deformation of Armco ingot iron at this temperature results in elongation of the crystals in the direction of straining. This deformation is accompanied by rupturing of some of the grain boundaries to form intercrystalline cracks. Swedish iron at the same temperature exhibits more marked grain distortion, as is shown in Fig. 71, but there is little indication of intercrystalline cracking.

At 450° C. the grain boundaries exposed at the surface are considerably widened, as in Fig. 87 (compare with Fig. 84, tested at room temperature). In Armco ingot iron, as distinct from the other materials tested, there is less difference between the structure at the surface of a specimen and that of its interior. At 450° C. widening of the grain boundaries on the surface is accompanied by intercrystalline cracking in the interior (Fig. 88).

In addition to the usual creep and tensile tests a special test on the material *A* was made to correlate the manner in which slip and grain boundary widening develop during the course of testing. The test was therefore conducted in stages and at each stage the specimen was examined microscopically. After re-

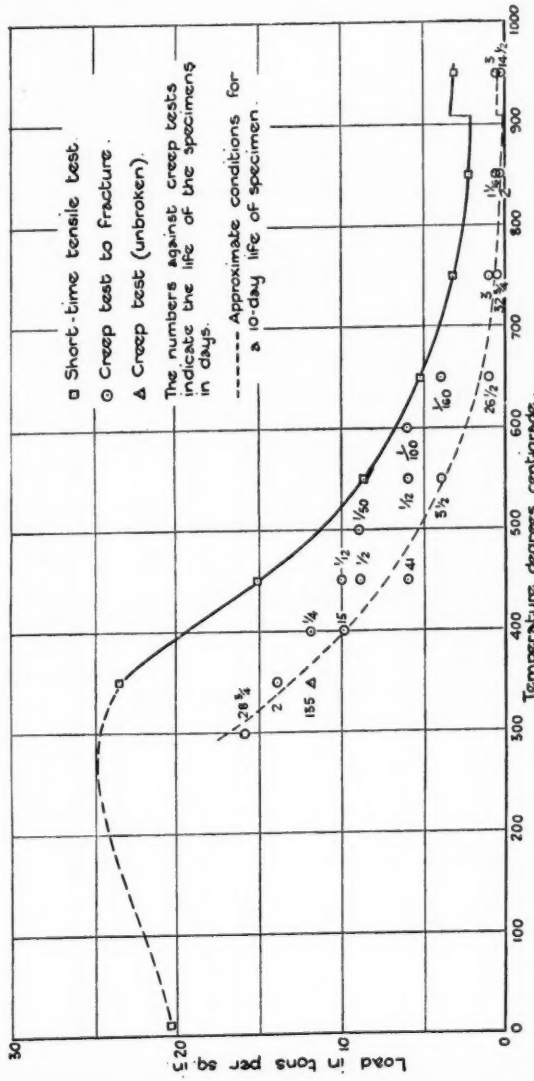

Fig. 12.—Load-Temperature Relationship. Armco ingot iron A.

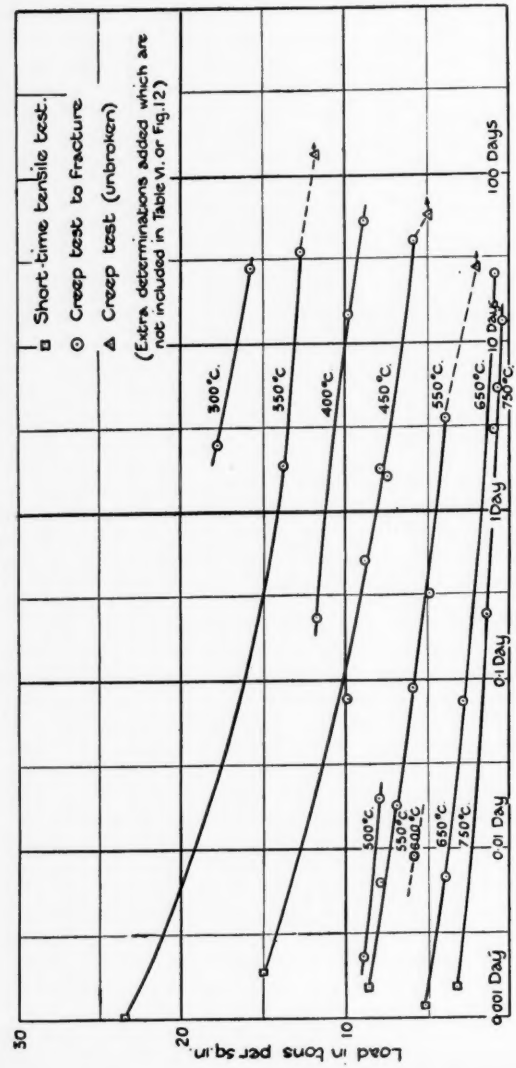

FIG. 13 (continued on Fig. 14).—Load-Life Relationship. Armco ingot iron *A*.

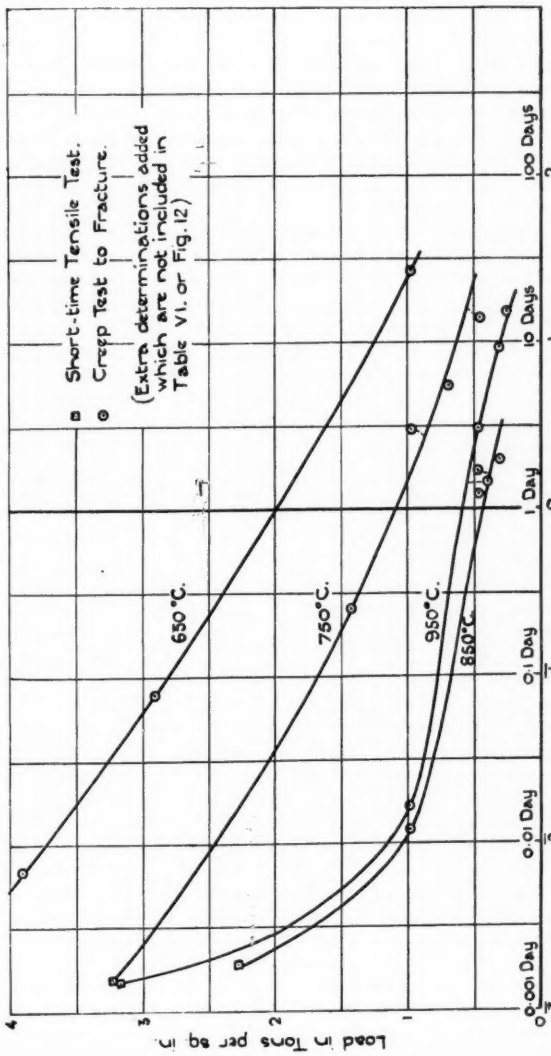

FIG. 14 (continued from Fig. 13).—Load-Life Relationship. Armco ingot iron A.

moving 0·025 in. by machining, the specimen was re-polished, re-etched and again loaded for a prescribed time. A nominal load of 7 tons per sq. in. was employed. The periodic microscopic examination showed that slip occurred during the first 54 hr. of testing, and no further development took place on the surface until the noticeable widening of the transverse grain boundaries occurred after 456 hr. A little slip also occurred at the conclusion of the test when the rate of deformation had considerably increased on account of the reduction in cross-sectional area and consequent increase of stress.

No recrystallisation of the ferrite crystals has been observed in specimens tested at 450° C. but marked recrystallisation occurred in specimen *A*80 (see Table VI.) after a test at 500° C. Again, at 550° C. marked recrystallisation is produced in the elongated crystal grains (Fig. 89). The recrystallisation has taken the form of sub-divisions whose grain boundaries are related to the planes of shear previously observed in the crystals (Fig. 85). This should be contrasted with the grain structure in Fig. 73 which shows the development of the " sub-boundary " type of recrystallisation in Swedish iron.

A specimen of material *H* tested at 550° C., also shows a marked development of new grains close to the fracture, and the new crystals formed are not dissimilar in shape to " sub-boundaries " (Fig. 90). After test at 600° C. material *A* shows marked recrystallisation at the fracture accompanied by rupturing of the former grain boundaries which are still discernible (Fig. 91).

The formation of intercrystalline cracking has been studied in numerous test specimens and there is a fairly clear-cut division into classes of behaviour ; first, tests in which the specimens have failed by transcrystalline fractures, and, secondly, tests in which the samples have failed by intercrystalline cracks. If in Fig. 12 a line is drawn through the three points whose co-ordinates are 10 tons per sq. in. at 400° C., 8½ tons per sq. in. at 500° C., and 6½ tons per sq. in. at 600° C., then all tests whose conditions are such as to fall below this line will show a definite intercrystalline weakness. At rapid rates of extension at temperatures from 650° to 850° C. the formation of intercrystalline cracks is possibly masked by recrystallisation effects.

The surface of a specimen of material *A* (Fig. 92) shows after a short-time tensile test at 650° C. fewer slip planes than at lower temperatures, but an increased degree of widening of the grain boundaries. The appearance of a longitudinal section of the same tensile test specimen (Fig. 93) is very similar to that of a creep test-piece which lasted only 10 min. at this temperature. In both cases the degree of elongation is high and marked recrystallisation and grain growth have occurred, the latter being indicated by the absence of smoothness of the grain boundaries.

There appears to be a little intercrystalline cracking even at these rapid rates of extension.

The rate of deformation plays an important part in the behaviour of the material at this temperature. At rapid rates the mode of deformation is as described above for the specimen illustrated in Fig. 93, but after testing to rupture at slow rates of extension other specimens show a small percentage elongation and fail by cracking at the grain boundaries with an almost complete absence of recrystallisation and grain growth (Fig. 94). Slow rates of extension result in the absence of visible slip on the surface (see Fig. 95). A still more marked influence of the rate of deformation at 850° C. is recorded in Figs. 96 and 97. In the short-time tensile test recrystallisation is almost complete, but no recrystallisation can be seen as a result of the creep test which lasted about two days.

In a creep test carried out at 950° C. no slip is observable on the surface (see Fig. 98), but the grains exhibit twinning which is characteristic of austenitic (γ) iron. The absence of slip in Swedish iron at the same temperature has already been recorded. The grain boundaries seem markedly widened by the extension, in this case 12 per cent., and the widened grain boundaries appear to be related to the intercrystalline cracking in the interior. A longitudinal section through another specimen (Fig. 99) shows that the grain size is coarser than that of the original material, but the discontinuities do not all lie in the grain boundaries. The presence of cracks and cavities in positions away from the grain boundaries may be due either to transcrystalline cracking or to the migration of the boundaries away from the cracks as they have been formed. Fig. 100 illustrates a crack penetrating a crystal which does not follow any grain boundary, but might have been related to a former grain boundary.

Attempts have been made to study the material by X-ray analysis. Some of the Armco ingot iron specimens considered above have been examined, together with specimens of cold-worked and annealed material of similar composition.

The examination did not reveal any marked preferred orientation set up as a result of creep, and the character and definition of the doublet indicated little permanent straining of the lattice. The differences in measurements of the parameter, which have been observed in specimens, appear to correspond to differences in volume which normally exist between cold-worked and annealed material. The changes of density on the cold working of iron are discussed by Adams.[6]

In connection with the difficulty of attack by the X-ray method, it should be noted that a strain of about $\frac{1}{300}$ is produced with a limit of proportionality as high as 18 tons per sq. in., assuming Young's modulus to be $E = 3{\cdot}0 \times 10^7$. The evidence of the

existence of such strain in a crystalline material has been obtained by Sachs and Weerts[7] in an investigation of elastic measurements by X-ray methods.

An X-ray examination of the surface of a ruptured specimen which had failed by intercrystalline fracture under a load of 0·44 tons per sq. in. at 850° C. has shown that the boundary surfaces consisted wholly of orientated iron crystals and exhibited no features which call for comment. There was no indication of material of a highly disorientated character such as has been associated with grain boundary material, but it is possible that after rupture rapid recrystallisation of the severed film might have take place. Such an event would destroy the disorientated character supposed to exist in material in the vicinity of the grain boundary.

Section 6.—The Behaviour of Carbonyl Iron *C*.

The composition and mechanical properties of the rolled plate of carbonyl iron are given in Tables I., II. and VII. It will be noted that the purity of this material exceeds 99·9 per cent. Swedish and carbonyl irons are alike in showing a considerable distortion of the ferrite crystals under deformation and in the manner of recrystallisation of the deformed ferrite,

Table VII.—*Tests on Carbonyl Iron* C in Vacuo.

Temperature ° C.	Specimen No. *C*.—	Machine Used.	Creep Load, or Ultimate Load on Amsler Machine. Tons per sq. in.	Elongation. %	Reduction of Area. %	Duration of Test. Days.	Duration of Test. Hr.	Micrograph See Fig.—
15	169	Amsler.*	18·2	47	94	..	..	101 and 102
300	118	Creep.	9·65	21	Unbroken	61	22	..
350	252	Amsler.	19·7	36	88	..	..	..
	221	Creep.	10·0	49†	91	..	2·2	103
400	115	Creep.	9·6	34†	96	..	0·25	104
	119	Creep.	5·63	59†	92	13	14	105 and 106
450	253	Amsler.	19·9	52	88	..	..	107 and 108
	168	Creep.	5·6	74†	90	..	3·25	109
500	170	Creep.	1·62	7	Unbroken	14	18	..
550	254	Amsler.	7·38	100	99	..	..	110
	256	Creep.	2·5	29†	67	8	19	..
	251	Creep.	1·5	7	Unbroken	21	0	..
650	257	Amsler.	3·84	109	99	..	..	111 and 112
	217	Creep.	0·66	3†	23	12	16	..
750	258	Amsler.	2·49	105	98	..	..	..
	255	Creep.	0·5	92	95	1	2	113
850	274	Amsler.	0·9	127	>99	..	..	114 and 115
	222	Creep.	0·35	100	97	..	1·75	..
950	275	Amsler.	2·51	146	>99	..	..	..
	220	Creep.	0·35	11	30	..	19	116
	218	Unstressed.	..	..	..	14	18	..
	219	Creep.	0·17	About 3	Unbroken	61	..	..

* Tested in air.
† Broke outside gauge length.

but carbonyl iron does not show any tendency to divide longitudinally into fibres as previously described. A curve showing the relation of load to temperature is plotted in Fig. 15 and the time required to produce rupture in a specimen after application of a load at a given temperature is plotted in Fig. 16.

Considerable necking occurs in carbonyl iron specimens tested at temperatures below 550° C., but at higher temperatures this form of localised deformation is replaced by a more general tapering of the parallel portion. Fractures of the cup and cone type are found only in short-time tensile test-pieces at temperatures up to 450° C. Creep tests at temperatures below 650° C. have resulted in irregular fractures. At higher temperatures the reductions of area are much greater than those recorded for other materials, and both creep and short-time tensile specimens exhibit chisel-edged fractures.

The microstructure before test is shown in Fig. 28, and is homogeneous and relatively free from non-metallic inclusions. The grain size, however, is somewhat variable, but there is no tendency for ferrite to deform locally in bands.

Fig. 101 (Plate XXXVI.) shows the degree of slip produced on the polished and etched surface of a test-piece after deformation at room temperature. There is no marked widening of the grain boundaries. The distortion of the crystals in the same test-piece (Fig. 102) appears to be more pronounced than in Swedish iron under similar conditions (Fig. 68), although the elongations of the two materials are not very different.

In a creep test carried out at 350° C. (Fig. 103) the crystals which are markedly elongated show the effect of the strain and are possibly beginning to recrystallise. Near the fracture, in addition to the wood-like appearance of the deformed crystals, strain bands occur at right angles to the direction of deformation. Similar strain bands have been observed in Armco ingot iron (Fig. 85) and have also been observed in rolled cupro-nickel by Adcock[8].

The interior of a specimen tested at a comparatively rapid rate at 400° C. is shown in Fig. 104. Near the fracture the crystals have been markedly deformed and twisted and are beginning to recrystallise. The rate of recrystallisation varies from one crystal to another, and similar tests on Swedish iron at the same temperature have shown the same phenomenon (see Fig. 71).

The surface of a creep specimen which ruptured after a test of 13 days at 400° C., exhibits well defined slip (Fig. 105). The interior of the same test-piece (Fig. 105) shows that the crystal grains are considerably drawn down and in process of recrystallisation. The appearance of these grains should be contrasted with those in Fig. 103. In this shorter test at a slightly lower temperature it is clear that the ferrite had only just commenced

to recrystallise. The test-pieces of carbonyl iron show considerable deformation in the form of necking, which is localised to such a degree that material at a distance of $\frac{1}{4}$ in. from the fracture may not even be appreciably deformed. Fig. 107 (Plate XXXVII.) shows an example of this absence of deformation $\frac{3}{8}$ in. from the fracture of a short-time tensile specimen tested at 450° C., whereas considerable deformation occurs close to the fracture of the same specimen, as in Fig. 108. Small grains of recrystallised ferrite have even appeared close to the fracture of a specimen subjected to a creep test which lasted only 3 hr. (Fig. 109).

In a short-time tensile test at 550° C. considerable recrystallisation has occurred close to the fracture (Fig. 110). Even $\frac{3}{8}$ in. from the fracture some recrystallisation has occurred, although the original ferrite grains have not been highly contorted but show marked elongation.

The localised deformation in a short-time tensile test at 650° C. has produced complete recrystallisation near the fracture (Fig. 111), whereas $\frac{3}{8}$ in. from the fracture the original grains are clearly defined. One of these original grains, as shown in Fig. 112 (Plate XXXVIII.), reveals the generation of new crystals like sub-boundaries within the main crystal. It is probable that the type of recrystallisation depicted in Fig. 111 has been preceded by the development of new grains in highly deformed ferrite, and the subsequent growth of the new grains. This mechanism is shown in its initial stages in Fig. 110, and is not similar to the process of sub-division of crystals in Fig. 112.

A longitudinal section through the fracture of a creep specimen tested at 750° C. is shown in Fig. 113. The test-piece has pulled down to a sharp pointed fracture and the material in this zone is completely recrystallised.

The surface of a short-time tensile specimen tested at 850° C. (Fig. 114), exhibits a relatively coarse formation of slip planes, and a section of this specimen $\frac{3}{8}$ in. from the fracture shows no marked deformation, although recrystallisation appears to be proceeding in one of the original crystals by a process of sub-division into smaller grains (Fig. 115).

In the tests at 950° C. the specimen has been deformed while in the austenitic (γ) state. General intercrystalline weakness does not appear, although discontinuities in the metal have been developed which appear to occupy the sites of present or former grain boundaries (Fig. 116).

Section 7.—Discussion of Results and Conclusions.

The properties of the materials considered in the present investigation are related to the following :

(1) The initial properties of the material.
(2) The effect of softening at high temperatures.

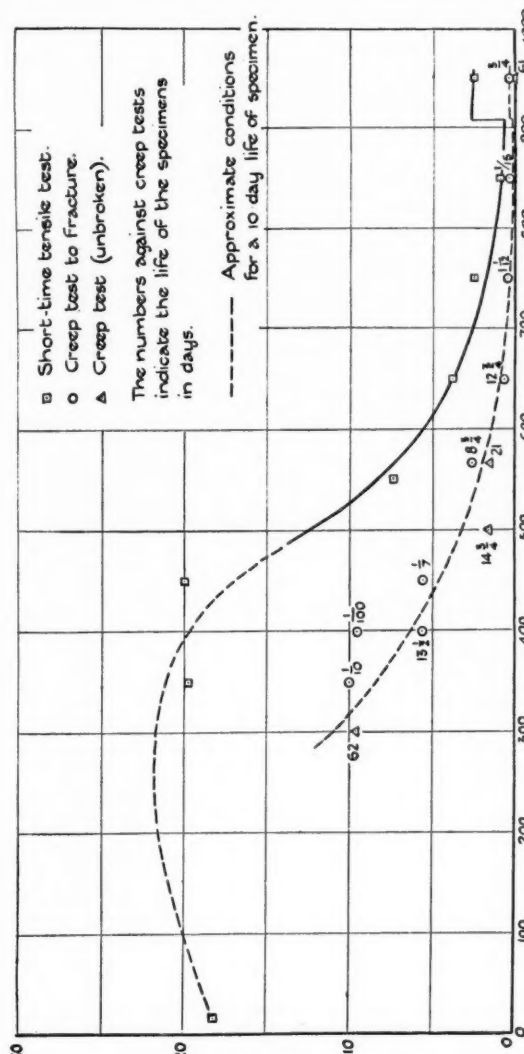

FIG. 15.—Load-Temperature Relationship. Carbonyl iron.

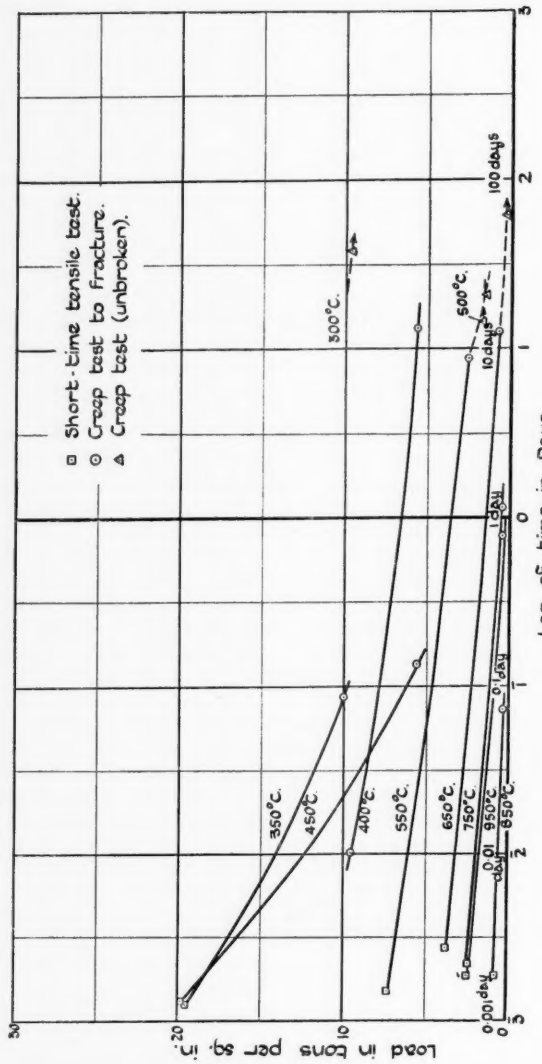

FIG. 16.—Load-Life Relationship. Carbonyl iron.

(3) The changes resulting from differences in chemical composition and mode of manufacture.
(4) The changes in the distribution of the chemical constituents, including the form of the carbide in mild steel and the distribution of oxygen.
(5) The changes in the physical structure of iron, which has a body-centred cubic lattice below the change point at approximately 900° C. and a face-centred cubic lattice at higher temperatures.

A summary of the principal forms of movement which can occur in tensile and creep tests has already been given in a previous paper[9] and is repeated below :

(1) Slip movement on the cleavage planes of the crystals and a general adjustment of crystal dimensions to accommodate the elongated crystals within the aggregate.
(2) General movement in the grain boundary material accompanied by auxiliary movement within the grains.
(3) Movement beginning by slip or grain boundary change, but proceeding after the initial period by contemporaneous recrystallisation of the strained material.

The tests on the seven materials between room temperature and 700° C. show that the extent to which changes produced by plastic flow are present is dependent upon the material, the temperature and time of test, and the degree of deformation. Many previous authors, such as Jeffries,[10] have assumed that metals are liable to intercrystalline breakdown under load at temperatures higher than that required to produce recrystallisation in " cold-worked " material. The latter has sometimes been termed the equicohesive temperature, but the use of this term does not allow for the very pronounced effect of the duration of annealing. In short-time tensile tests the maximum value of the ultimate load occurs at a temperature of approximately 250° C. The softening range, as determined by the recrystallisation of " cold-worked " material, is probably in the vicinity of 450° C. to 600° C.[11] but in creep tests this range of temperature is lowered to 350° C. to 500° C. as softening by heat-treatment occurs more rapidly when accompanied by simultaneous deformation. The behaviour of any " cold-worked " metal is greatly influenced by the duration of treatment at suitable recrystallisation temperatures. If " cold-worked " material is reheated in a certain low temperature range sensible ageing occurs and hardening is developed. In view of the possibility of hardening after cold-work it is clear that the relationship of softening to heat-treatment cannot be direct. The creep rate in the third stage of a test, as described later, is determined by the rate of softening.

The variation in the rate of deformation during a creep test is an important factor in the consideration of the mode of deformation. Curves expressing the relationship between deformation and time consist of four distinct stages : First, the initial strain on loading ; secondly, a fairly rapid deformation whose rate decreases with time ; thirdly, a period of steady creep at a minimum rate of strain ; and finally, a period of increasing

rate of strain, which leads to fracture. For carbon steels between 300° C. and 600° C. under varying stresses producing the *same minimum rates* of creep, it has been shown[12] that the degree of deformation in the first and second stages is large at low temperatures but decreases to a small value at 600° C. The length of the third period of steady creep is not greatly altered by change in temperature, but the duration of the fourth period increases with the temperature of testing.

In a stage test on Armco ingot iron at 450° C., it is found that slip is the principal form of movement during the first two stages of deformation mentioned above. In the third stage movement occurs without appreciable change on the surface, whereas in the fourth stage the higher stress, resulting from the reduction in the cross-sectional area of the specimen, gives an increased rate of deformation leading to failure. Additional slip bands are observable in this stage. It is probable that in the third stage movement occurs by continuous recrystallisation and the rate of hardening from strain is equal to the rate of softening from annealing.

Slip is most obvious in carbonyl iron and Swedish iron and is least visible in Armco ingot iron and mild steel. The principal visible form of movement in these materials below 300° C. is by slip, which diminishes with increasing temperature up to 850° C.

At low temperatures, such as 450° C., initial deformation of a polycrystalline aggregate occurs by slip. Re-bonding occurs in a new position, but the process is not without a marked effect on the atoms adjoining the slip plane, as is shown in the permanent alteration of the properties of the material. Prolonged exposure to the temperature of the creep test, no doubt, produces early and invisible forms of crystallisation which hamper further slip movement by a spreading of the intensely disturbed zone of atoms which has been formed as a result of slip. The results show that the distribution of movement in a polycrystalline aggregate severely deformed by creep and short-time tensile tests is far from being uniform, and "strained areas" are developed which can be detected by sectioning, polishing, and etching methods. These "strained areas" after prolonged exposure to a suitable testing temperature form the place of origin of new crystals.

X-ray examination suggests that the lattice of a crystal grain can be strained to a degree corresponding to the elastic limit. Beyond this point slip occurs and the stress is relieved to a value possibly slightly less than that at the point of slipping. In order that slip may continue the stress must again build up to a value as great as or greater than the previous elastic limit. Slip is not a continuous movement, but occurs in a series of

steps which correspond to lateral movements on groups of parallel planes. In terms of the atomic size this movement is appreciable, for the blocks have dimensions corresponding to several thousands of atoms. This type of movement[13] was found in creep tests on Staybrite steel and phosphor bronze which were tested, on account of the relatively low testing temperatures, under loads which were high in relation to the yield points of the materials. Although there was no appreciable deformation for periods of many hours, a marked and apparently instantaneous extension occurred suddenly as a result of slight vibrations in the room. The application of load was continued without appreciable elongation until a further sudden yielding took place.

The present work shows that deformation at high temperatures results in recrystallisation, and the term is regarded as including any process which results in crystalline recovery. As far as its ultimate effect is concerned, the subject of recrystallisation is divisible into three sections :

(*a*) Grain growth.
(*b*) Generation of new crystals by *full* recrystallisation.
(*c*) Division of existing crystals into smaller units : "Sub-crystallisation."

Thus deformation by creep tends to produce fragmentation of the crystals, whereas the result of prolonged treatment at an annealing temperature results in more or less complete crystalline recovery. In some cases less recrystallisation is found at high temperatures in a longer time of test than in a shorter one, while the converse is true at lower temperatures of test. In the former case the crystals tend to sub-divide in a manner similar to the formation of sub-boundaries, while in the latter full recrystallisation occurs.

In the case of slight deformation at an intermediate range of temperature re-orientation effects are set up in the strained crystals without the formation of new grains. The phenomenon is somewhat analogous to the results of re-annealing deformed material, in that lightly deformed material does not recrystallise, but undergoes re-orientation within the crystals, and consequently considerable re-adjustment of the internal grain boundaries is necessary to produce a stable system. Grain growth, therefore, occurs on the part of favourably orientated crystals, and such grain growth is found in some of the test-pieces in the present series.

The normal manner of failure in test-pieces of iron and mild steel fractured below 400° C. is by transcrystalline rupture. In longer time creep tests at 400° C. Armco ingot iron exhibits intercrystalline cracking, but no characteristic intercrystalline failure is found in the other materials tested at this temperature. Transcrystalline rupture probably results from the fact that the centre of a crystal, owing to geometrical conditions, deforms more

easily than the material in the region of the grain boundaries. Transcrystalline rupture is an indefinite term and includes all types of failure which are not definitely intercrystalline.

In addition to slip, other forms of movement can be seen on the surface of test-pieces. At temperatures higher than about 500° C., varying with different materials, there is on the surface a widening of transverse grain boundaries* which may develop into intercrystalline cracks for the depth of a crystal or so. These cracks show little tendency to spread into the material except in Armco ingot iron. The tests on Armco ingot iron also show that intercrystalline cracks may develop by internal separation of the grains.

Failure by intercrystalline cracking is found in materials subjected to internal strain accompanied by corrosive conditions, and in alloy steels fractured by prolonged application of stress at elevated temperatures. In the latter case the crystals have been stiffened by alloying additions,[14] but this does not appear to strengthen the grain boundaries.

The present work suggests that intercrystalline cracking occurs in a material which does not accommodate its localised differences in stress by recrystallisation. The facility of recrystallisation prevents a material from becoming unduly work-hardened and does, in general, hinder the formation of cracks which lead to rupture. In creep tests at high temperatures Armco ingot iron is the material which is most prone to failure by intercrystalline cracking. The particular behaviour of this material is probably connected with the distribution of oxygen, nitrogen and other constituents, which leads to a stiffening action on the ferrite without appreciable strengthening of the grain boundaries.

The mode of failure of mild steel at high temperature is not ascribable to a general intercrystalline form of cracking, although a form of weakness develops which leads to the formation of cavities in the contact zone of ferrite and pearlite grains. At temperatures below 300° C. the presence of the harder constituent, pearlite, probably has a marked stiffening effect on the ferrite grains, and this effect diminishes with rise of temperature up to the A_1 point.

Apart from the mode of rupture and degree of recrystallisation, spheroidisation of the carbide in mild steel is found to take place. After long duration tests over the range of temperature 450° C.–650° C., and after short-time tensile tests at 550° C. and 650° C., the pearlite present in mild steel begins to spheroidise under the action of the temperature and deformation. If spheroidisation is continued to completion, the iron carbide is aggregated

* See discussion to paper by D. Hanson and M. A. Wheeler, *Journal of the Institute of Metals*, 1931, No. 1, p. 229.

MATERIALS BEFORE TEST.

FIG. 17.—Mild Steel *NGK*. Sulphur Prints. Upper, longitudinal section; lower, transverse. × 1.

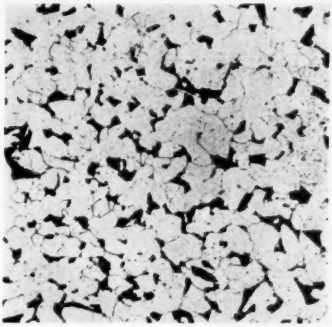

FIG. 18.—Mild Steel *NGK*. Microstructure. × 150.

FIG. 19.—Mild Steel *M*. Microstructure. × 150.

FIG. 20.—Swedish Iron *S*. Phosphorus etch prints. Upper, longitudinal section; lower, transverse. × 1.

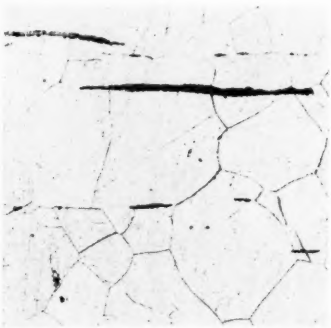

FIG. 21.—Swedish Iron *S*. Microstructure. × 150.

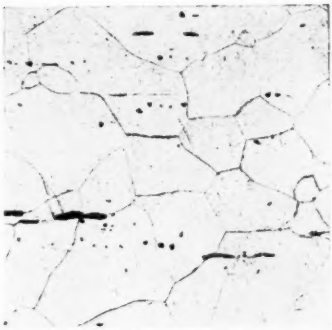

FIG. 22.—Swedish Iron *L*. Microstructure. × 150.

(Micrographs reduced to two-thirds linear in reproduction.)

[*Jenkins & Mellor.*
[*To face p.* 224.

MATERIALS BEFORE TEST.

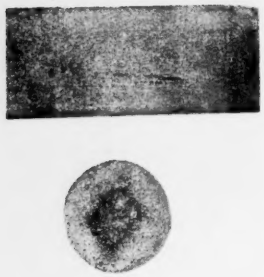

FIG. 23.—Armco Ingot Iron *A*. Sulphur Prints. Upper, longitudinal section; lower, transverse. × 1.

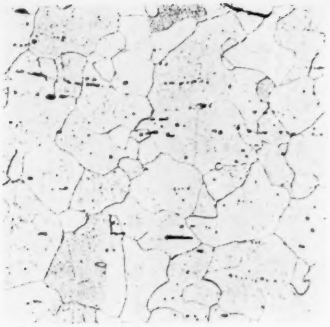

FIG. 24.—Armco Ingot Iron *A*. Microstructure. × 150.

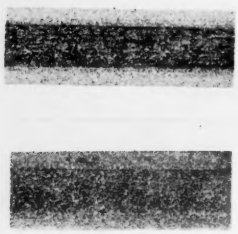

FIG. 25.—Armco Ingot Iron *H*. Upper, sulphur print; lower, phosphorus etch print. × 1.

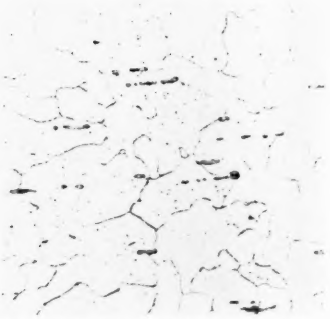

FIG. 26.—Armco Ingot Iron *H*. Microstructure. × 150.

FIG. 27.—Carbonyl Iron *C*. Upper, sulphur print; lower, phosphorus etch print.

FIG. 28.—Carbonyl Iron *C*. Microstructure. × 150.

(Micrographs reduced to two-thirds linear in reproduction.)

[Jenkins & Mellor.

MILD STEEL.

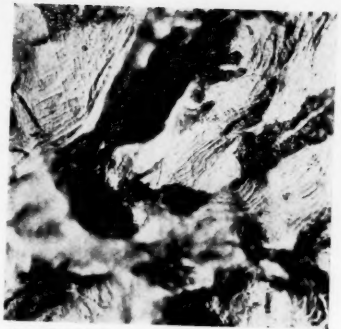

FIG. 29.—15° C. 25·4 tons per sq. in. Short-time tensile test. *NGK*. 271. Surface, $\frac{3}{8}$ in. from fracture. × 500.

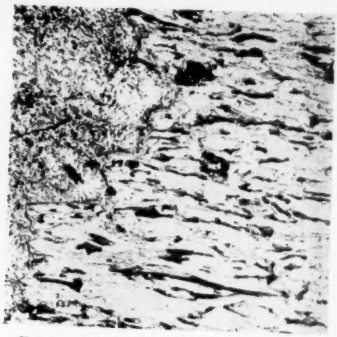

FIG. 30.—15° C. 25·4 tons per sq. in. Short-time tensile test. *NGK*. 271. Section at iron-plated fracture. × 150.

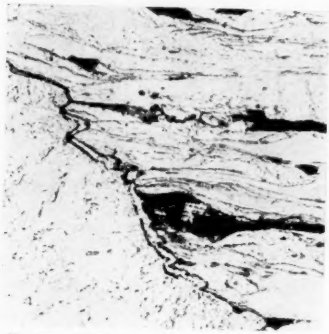

FIG. 31.—450° C. 21·3 tons per sq. in. Short-time tensile test. *NGK*. 240. Section at iron-plated fracture. × 500.

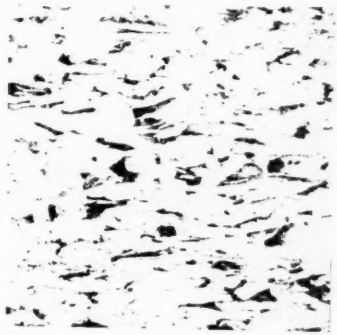

FIG. 32.—450° C. 15·8 tons per sq. in. Creep test. 2 days 1 hr. *NGK*. 236. Section $\frac{1}{4}$ in. from fracture. × 150.

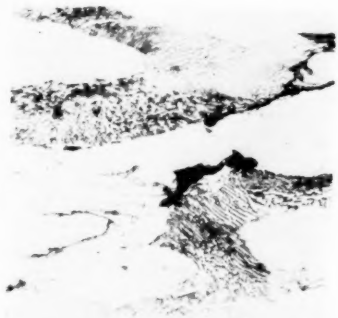

FIG. 33.—450° C. 13·25 tons per sq. in. Creep test. 12 days 22 hr. *NGK*. 280. Section close to fracture. × 1,500.

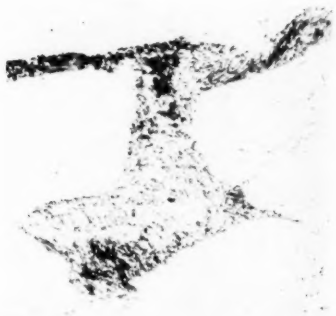

FIG. 34.—450° C. 12 tons per sq. in. Creep test. 21¼ days. *NGK*. 272. Section close to fracture. × 1,500.

(Micrographs reduced to two-thirds linear in reproduction.)

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FIG. 35.—450° C. 12 tons per sq. in. Creep test. 21½ days. *NGK*. 272. Section ⅛ in. from fracture. × 1,500.

FIG. 36.—450° C. 21½ days. *NGK*. 272*X*. Unstressed material. × 1,500.

FIG. 37.—500° C. 10 tons per sq. in. Creep test. 2 days 18 hr. *NGK*. 268. Section ⅛ in. from fracture. × 500.

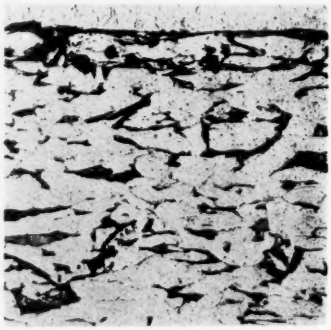

FIG. 38.—500° C. 9·9 tons per sq. in. Creep test. 4 days 14 hr. *NGK*. 223. Edge of iron-plated section ⅛ in. from fracture. × 150.

FIG. 39.—500° C. 9·9 tons per sq. in. Creep test. 7 days 7 hr. *NGK*. 238. Section close to fracture. × 500.

(Micrographs reduced to two-thirds linear in reproduction.)

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FIG. 40.—550° C. 13·7 tons per sq. in. Short-time tensile test. *NGK*. 235. Surface ½ in. from fracture. × 500.

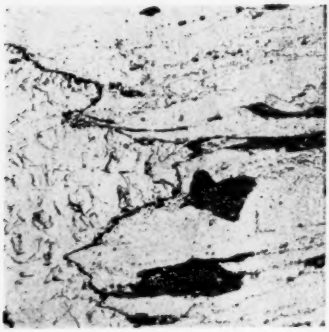

FIG. 41.—550° C. 13·7 tons per sq. in. Short-time tensile test. *NGK*. 235. Section at iron-plated fracture. × 500.

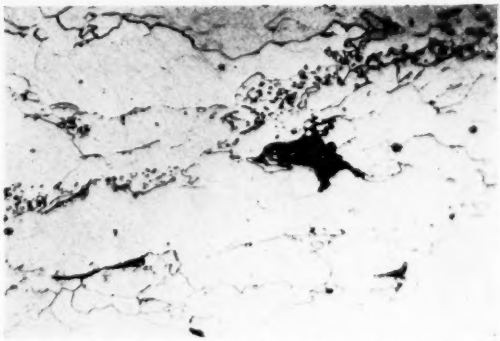

FIG. 42.—550° C. 4 tons per sq. in. Creep test. 24 days. *NGK*. 281. Section close to fracture. × 1,500.

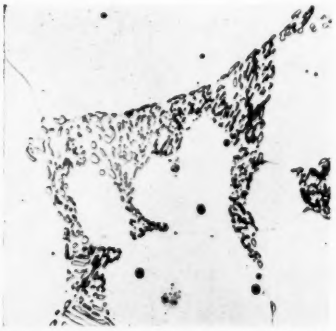

FIG. 43.—550° C. 24 days. *NGK*. 281*X*. Unstressed material. × 1,500.

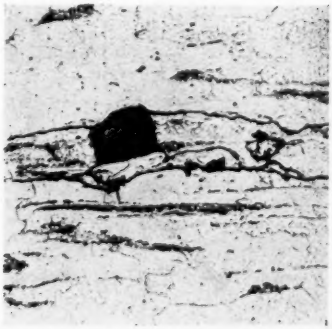

FIG. 44.—650° C. 8·24 tons per sq. in. Short-time tensile test. *NGK*. 237. Section $\frac{1}{16}$ in. from fracture. × 500.

(Micrographs reduced to two-thirds linear in reproduction.)

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MILD STEEL.

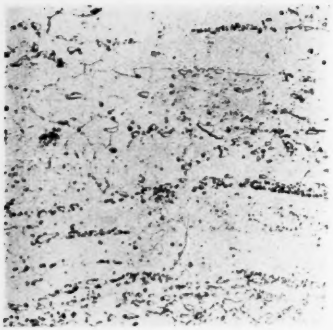

FIG. 45.—650° C. 1·94 tons per sq. in. Creep test. 6 days. *NGK*. 226. Section $\frac{1}{4}$ in. from fracture. × 500.

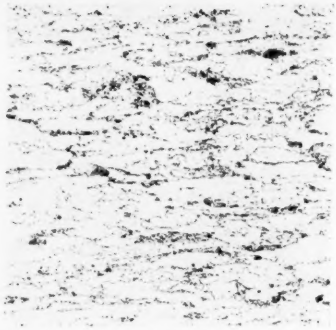

FIG. 46.—750° C. 4·64 tons per sq. in. Short-time tensile test. *NGK*. 242. Section $\frac{3}{8}$ in. from fracture. × 150.

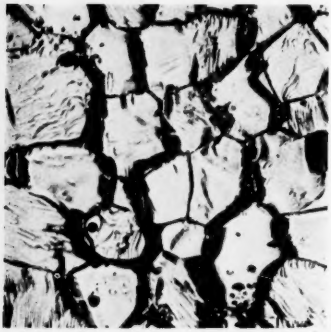

FIG. 47.—850° C. 0·5 ton per sq. in. Creep test. 31$\frac{1}{2}$ days. *NGK*. 246. Surface 1 in. from fracture. × 300.

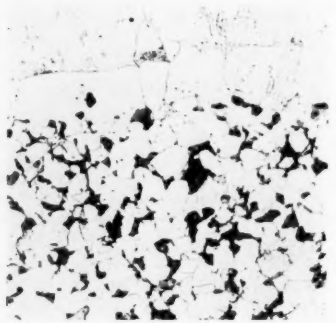

FIG. 48.—850° C. 0·5 ton per sq. in. Creep test. 31$\frac{1}{2}$ days. *NGK*. 246. Edge of section $\frac{1}{8}$ in. from fracture. × 150.

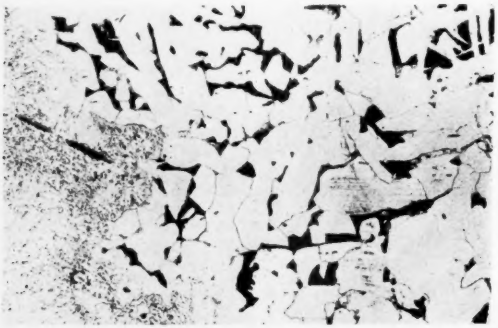

FIG. 49.—950° C. 0·5 ton per sq. in. Creep test. 12 days 4 hr. *NGK*. 270. Section at iron-plated fracture. × 150.

(Micrographs reduced to two-thirds linear in reproduction.)

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MILD STEEL.

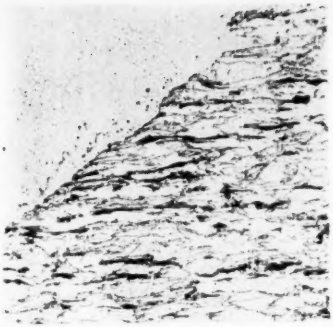

FIG. 50.—450° C. 17·8 tons per sq. in. Creep test. 18 days. *M.* 71. Section at copper-plated fracture. × 150.

FIG. 51.—450° C. 17·8 tons per sq. in. Creep test. 18 days. *M.* 71. Section $\frac{1}{8}$ in. from fracture. × 1,500.

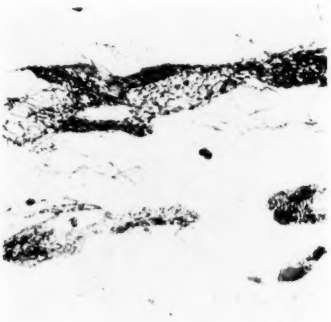

FIG. 52.—450° C. 17·8 tons per sq. in. Creep test. 18 days. *M.* 71. Section close to fracture. × 1,500.

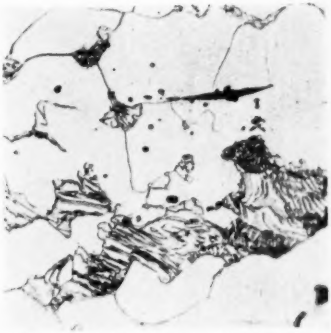

FIG. 53.—450° C. 18 days. Unstressed end of test-piece. *M.* 71. × 1,500.

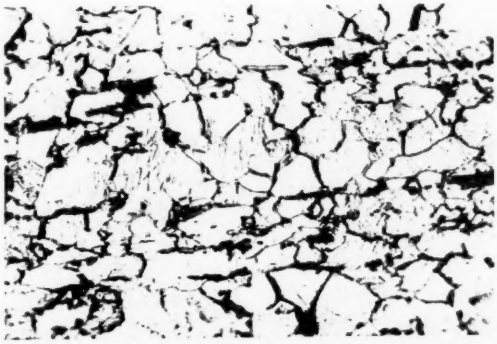

FIG. 54.—450° C. 17·8 tons per sq. in. Creep test. 7 days. *M.* 124. Surface $\frac{1}{4}$ in. from fracture. × 300.

(Micrographs reduced to two-thirds linear in reproduction.)

[*Jenkins & Mellor.*

MILD STEEL.

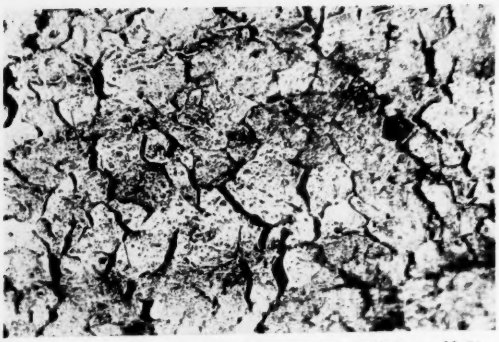

FIG. 55.—500° C. 9·9 tons per sq. in. Creep test. 17½ days. *M.* 74. Surface after test. × 300.

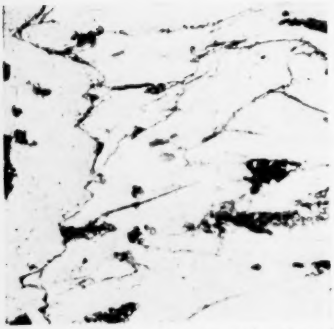

FIG. 56.—500° C. 9·9 tons per sq. in. Creep test. 17½ days. *M.* 74. Section at copper-plated fracture. × 1,500.

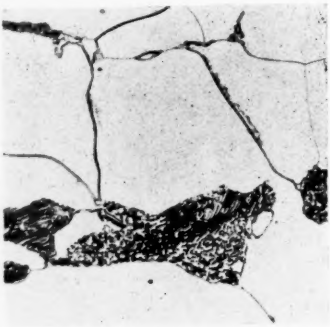

FIG. 57.—500° C. 8·9 tons per sq. in. Creep test. 34 days. Unbroken. *M.* 95. Section after test. × 1,500.

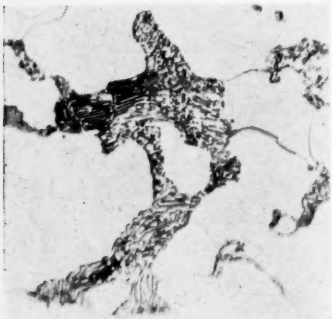

FIG. 58.—500° C. 34 days. Unstressed end of test-piece. *M.* 95. × 1,500.

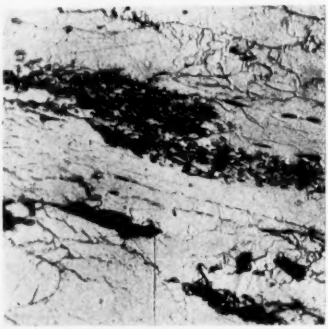

FIG. 59.—550° C. 8·2 tons per sq. in. Creep test. 21½ hr. *M.* 61. Section after test. × 1,500.

(Micrographs reduced to two-thirds linear in reproduction.)

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MILD STEEL.

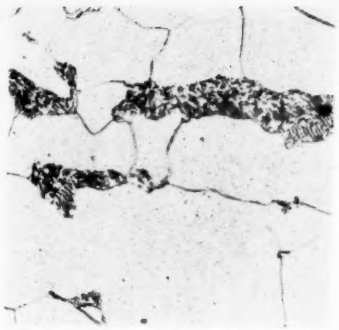

FIG. 60.—550° C. 43 days. Unstressed end of test-piece. *M.* 76. × 1,500.

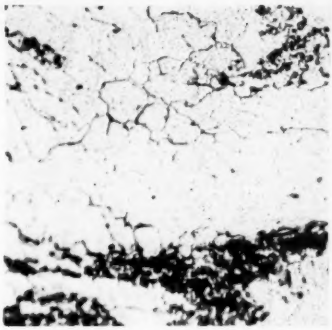

FIG. 61.—650° C. 5 tons per sq. in. Creep test. 1 hr. 37 min. *M.* 77. Section close to fracture. × 1,500.

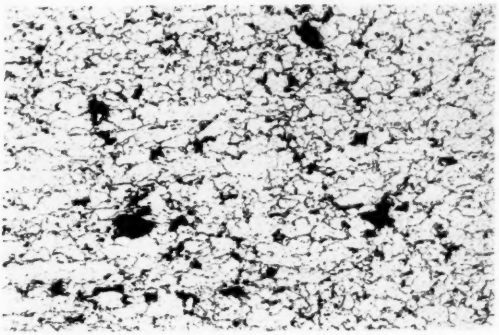

FIG. 62.—650° C. 1·46 tons per sq. in. Creep test. 34 days. *M.* 117. Section $\frac{1}{8}$ in. from fracture. × 150.

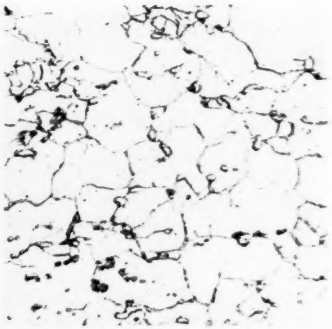

FIG. 63.—650° C. 1·46 tons per sq. in. Creep test. 34 days. *M.* 117. Section $\frac{1}{8}$ in. from fracture. × 500.

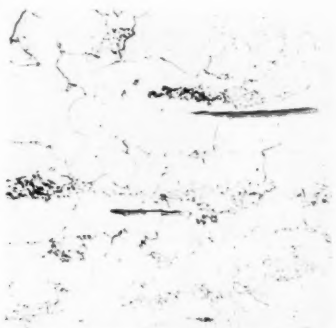

FIG. 64.—650° C. 34 days. Unstressed end of test-piece. *M.* 117. × 500.

(Micrographs reduced to two-thirds linear in reproduction.)

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MILD STEEL AND SWEDISH IRON.

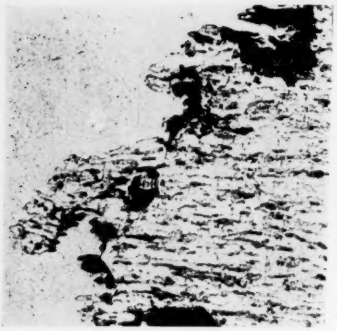

FIG. 65.—750° C. 2·9 tons per sq. in. Creep test. 23 min. *M.* 75. Section at copper-plated fracture. × 150.

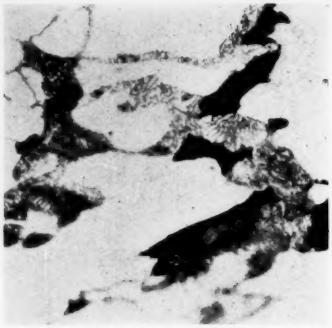

FIG. 66.—750° C. 0·94 ton per sq. in. Creep test. 3 days 15 hr. *M.* 116. Section $\frac{1}{4}$ in. from fracture. × 1,500.

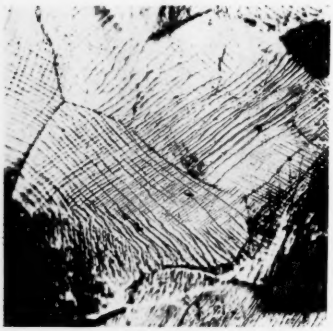

FIG. 67.—15° C. 19·7 tons per sq. in. Short-time tensile test. *S.* 145. Surface $\frac{1}{4}$ in. from fracture. × 300.

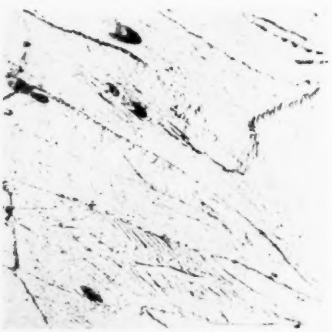

FIG. 68.—15° C. 19·7 tons per sq. in. Short-time tensile test. *S.* 145. Section $\frac{1}{16}$ in. from fracture. × 500.

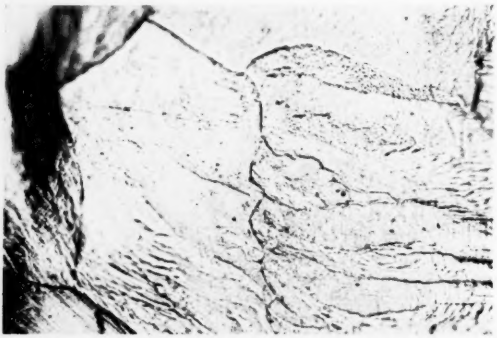

FIG. 69.—400° C. 11·8 tons per sq. in. Creep test. 27 hr. 20 min. *S.* 107. Surface $\frac{3}{4}$ in. from fracture. × 500.

*(Micrographs reduced to two-thirds linear in reproduction.)

Jenkins & Mellor

SWEDISH IRON.

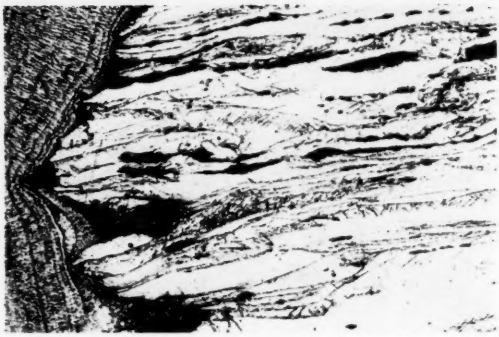

FIG. 70.—400° C. 11·8 tons per sq. in. Creep test. 27 hr. 20 min. *S.* 107. Section at iron-plated fracture. × 150.

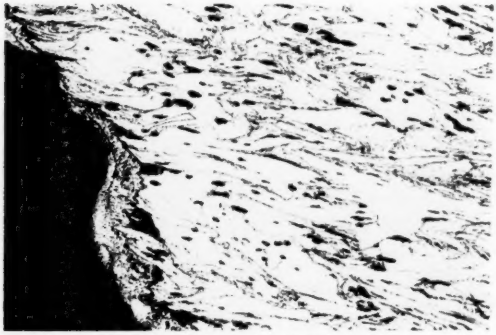

FIG. 71.—400° C. 9·9 tons per sq. in. Creep test. 4 days. *L.* 125. Section at iron-plated fracture. × 150.

FIG. 72.—450° C. 7·9 tons per sq. in. Creep test. 2 days 10 hr. *S.* 72. Surface after test. × 300.

(Micrographs reduced to two-thirds linear in reproduction.)

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SWEDISH IRON.

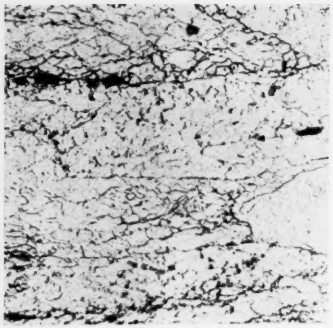

FIG. 73.—450° C. 5·8 tons per sq. in. Creep test. 26 days. *S.* 92. Section $\frac{1}{2}$ in. from fracture. × 500.

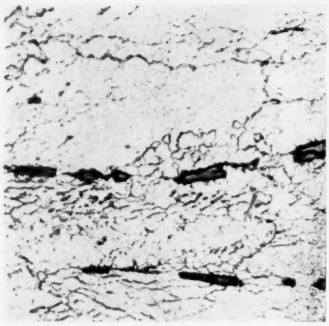

FIG. 74.—500° C. 3·9 tons per sq. in. Creep test. 14 days 6 hr. *L.* 108. Section $\frac{1}{8}$ in. from fracture. × 500.

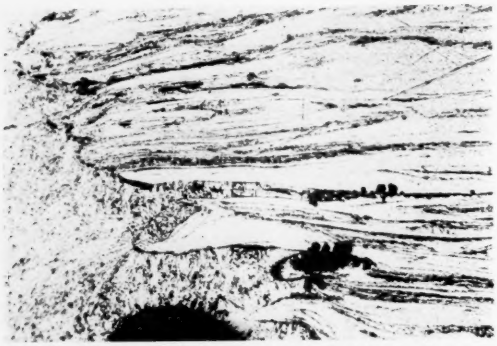

FIG. 75.—550° C. 10·45 tons per sq. in. Short-time tensile test. *S.* 147. Section at iron-plated fracture. × 150.

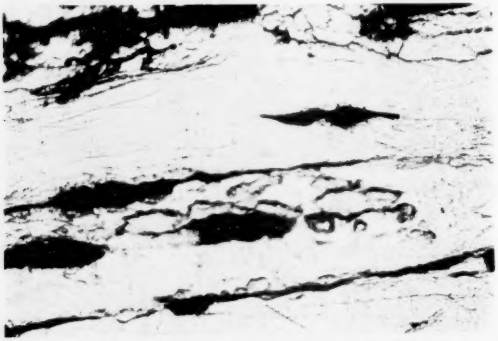

FIG. 76.—550° C. 3·9 tons per sq. in. Creep test. 18 hr. *S.* 69. Section close to fracture. × 1,500.

(Micrographs reduced to two-thirds linear in reproduction.)

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SWEDISH IRON.

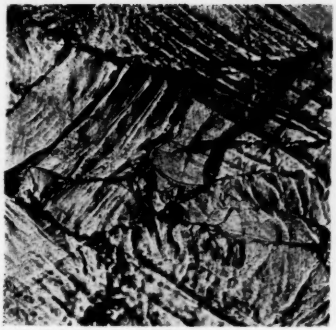

FIG. 77.—550° C. 2·9 tons per sq. in. Creep test. 8 days 2 hr. *S.* 78. Surface 1 in. from fracture. × 300.

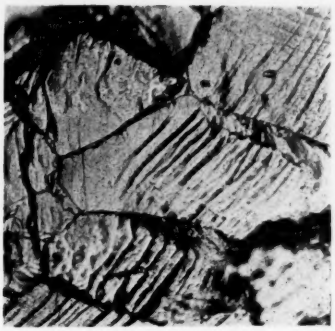

FIG. 78.—650° C. 2·9 tons per sq. in. Creep test. 27 min. *S.* 79. Surface after test. × 300.

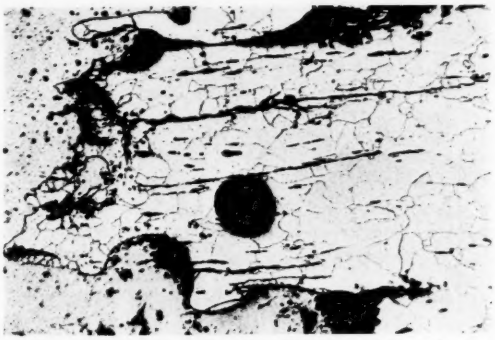

FIG. 79.—650° C. 2·9 tons per sq. in. Creep test. 27 min. *S.* 79. Section at copper-plated fracture. × 150.

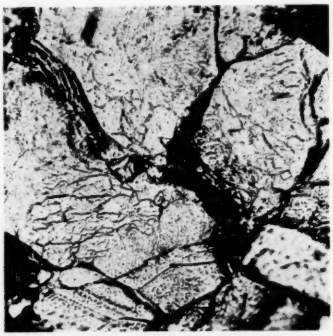

FIG. 80.—650° C. 0·94 ton per sq. in. Creep test. 28 days. *S.* 102. Surface $\frac{1}{4}$ in. from fracture. × 300.

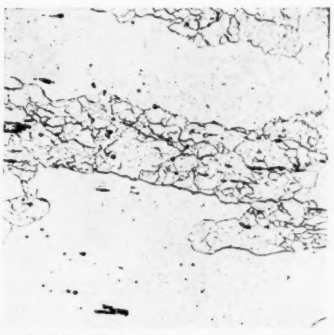

FIG. 81.—850° C. 1·55 tons per sq. in. Short-time tensile test. *S.* 150. Section $\frac{3}{4}$ in. from fracture. × 150.

(Micrographs reduced to two-thirds linear in reproduction.)

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SWEDISH AND INGOT IRONS.

FIG. 82.—850° C. 0·3 ton per sq. in. Creep test. 1 day 22 hr. *S.* 109. Surface 1¼ in. from fracture. × 150.

FIG. 83.—950° C. 0·5 ton per sq. in. Creep test. 8 days 5 hr. *S.* 231. Edge of iron-plated section ⅛ in. from fracture. × 150.

FIG. 84.—15° C. 18·5 tons per sq. in. Short-time tensile test. *H.* 173. Surface ½ in. from fracture. × 500.

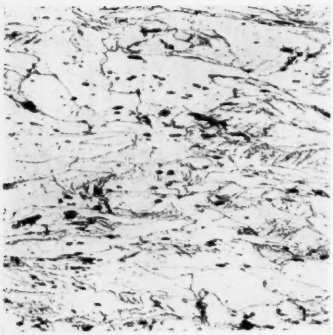

FIG. 85.—350° C. 23·5 tons per sq. in. Short-time tensile test. *A.* 163. Section ☆ in. from fracture. × 150.

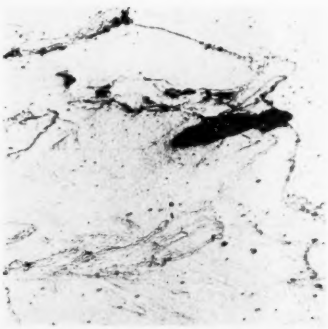

FIG. 86.—400° C. 9·9 tons per sq. in. Creep test. 23 hr. *H.* 172. Section at iron-plated fracture. × 500.

(Micrographs reduced to two-thirds linear in reproduction.)

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INGOT IRON.

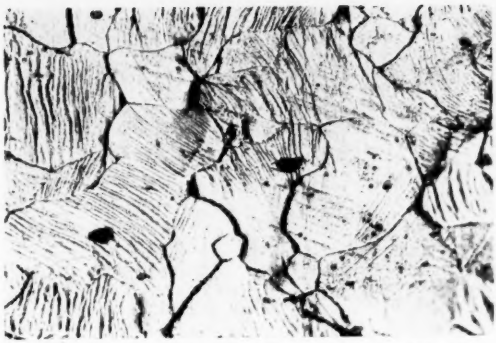

FIG. 87.—450° C. 7·5 tons per sq. in. Creep test. 1 day 16 hr. *A*. 162. Surface 1 in. from fracture. × 300.

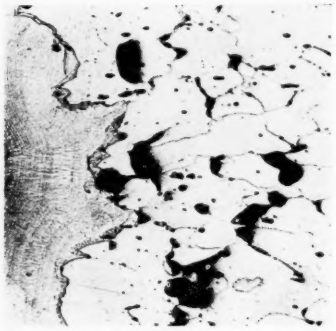

FIG. 88.—450° C. 7·5 tons per sq. in. Creep test. 1 day 16 hr. *A*. 162. Section at iron-plated fracture. × 150.

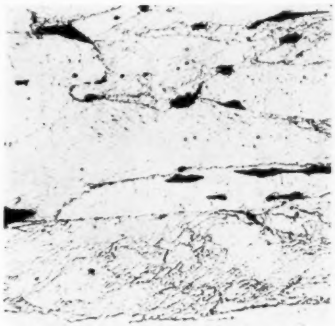

FIG. 89.—550° C. 8·62 tons per sq. in. Short-time tensile test. *A*. 165. Section $\frac{1}{4}$ in. from fracture. × 500.

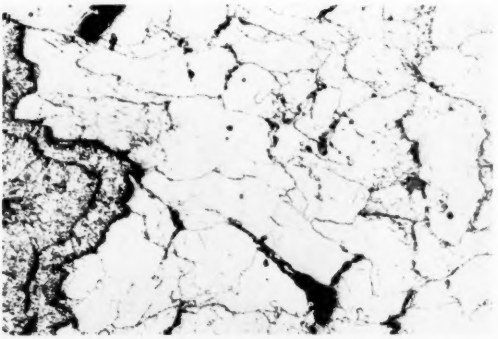

FIG. 90.—550° C. 3·9 tons per sq. in. Creep test. 2 days 16 hr. *H*. 174. Section at iron-plated fracture. × 150.

(Micrographs reduced to two-thirds linear in reproduction.)

INGOT IRON.

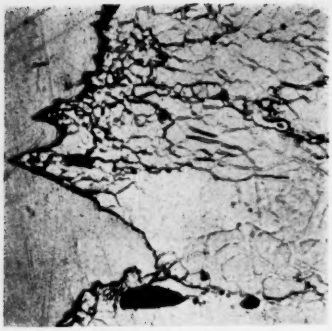

FIG. 91.—600° C. 5·94 tons per sq. in. Creep test. 13 min. *A*. 82. Section at copper-plated fracture. × 1,500.

FIG. 92.—650° C. 5·2 tons per sq. in. Short-time tensile test. *A*. 166. Surface ½ in. from fracture. × 300.

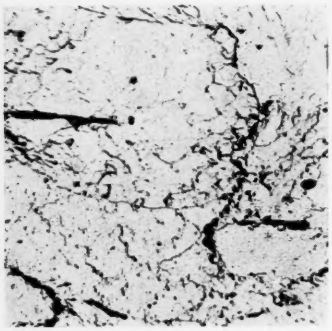

FIG. 93.—650° C. 5·2 tons per sq. in. Short-time tensile test. *A*. 166. Section ¼ in. from fracture. × 500.

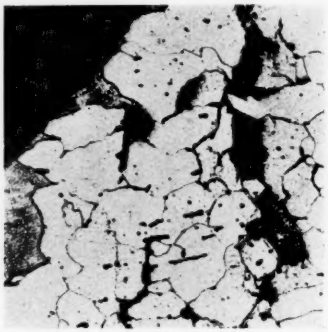

FIG. 94.—650° C. 0·94 ton per sq. in. Creep test. 26½ days. *A*. 28. Section at iron-plated fracture. × 150.

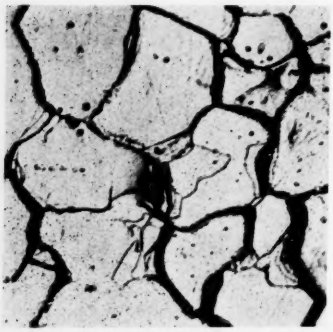

FIG. 95.—750° C. 0·44 ton per sq. in. Creep test. 33 days. *A*. 171. Surface ¼ in. from fracture. × 800.

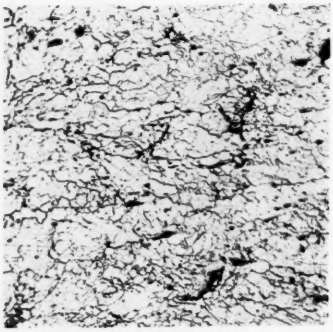

FIG. 96.—850° C. 2·26 tons per sq. in. Short-time tensile test. *A*. 157. Section ¼ in. from fracture. × 150.

(Micrographs reduced to two-thirds linear in reproduction.)

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INGOT IRON.

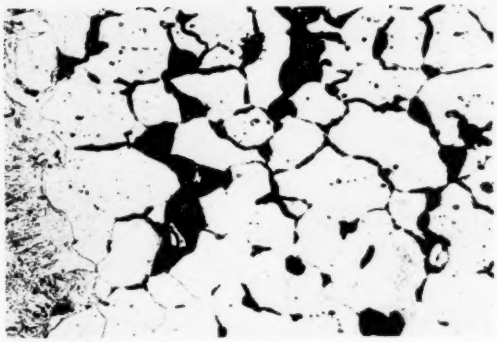

FIG. 97.—850° C. 0·29 ton per sq. in. Creep test. 1 day 22 hr. *A*. 93. Section at iron-plated fracture. × 150.

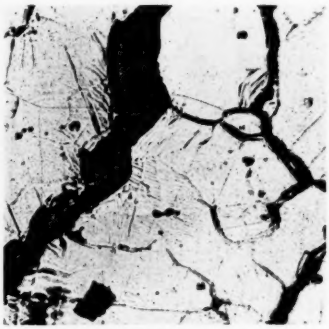

FIG. 98.—950° C. 0·38 ton per sq. in. Creep test. 1 day 12 hr. *A*. 167. Surface after test $\frac{1}{4}$ in. from fracture. × 300.

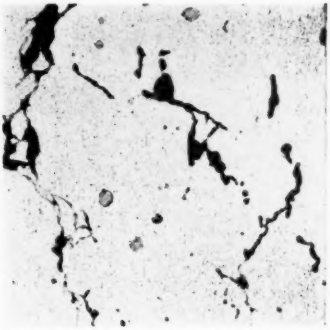

FIG. 99.—950° C. 0·44 ton per sq. in. Creep test. 3 days 3 hr. *H*. 175. Section $\frac{1}{16}$ in. from fracture. × 150.

FIG. 100.—950° C. 0·24 ton per sq. in. Creep test. 14$\frac{1}{2}$ days. *A*. 90. Edge of iron-plated section $\frac{1}{4}$ in. from fracture. × 500.

(Micrographs reduced to two-thirds linear in reproduction.)

Jenkins & Mellor.

CARBONYL IRON.

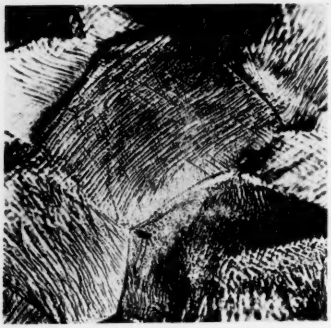

FIG. 101.—15° C. 18·2 tons per sq. in. Short-time tensile test. *C*. 169. Surface $\frac{1}{2}$ in. from fracture. × 300.

FIG. 102.—15° C. 18·2 tons per sq. in. Short-time tensile test. *C*. 169. Section $\frac{1}{16}$ in. from fracture. × 150.

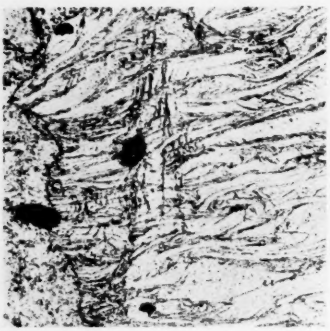

FIG. 103.—350° C. 10 tons per sq. in. Creep test. 2 hr. 10 min. *C*. 221. Section at iron-plated fracture. × 150.

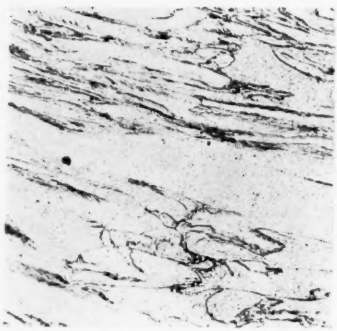

FIG. 104.—460° C. 9·6 tons per sq. in. Creep test. 15 min. *C*. 115. Section $\frac{1}{16}$ in. from fracture. × 150.

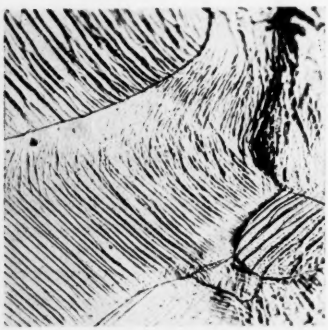

FIG. 105.—400° C. 5·63 tons per sq. in. Creep test. 13 days 15 hr. *C*. 119. Surface $1\frac{1}{4}$ in. from fracture. × 300.

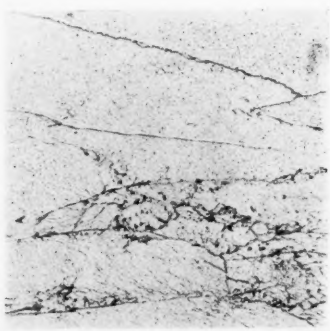

FIG. 106.—400° C. 5·63 tons per sq. in. Creep test. 13 days 15 hr. *C*. 119. Section $\frac{1}{8}$ in. from fracture. × 500.

(Micrographs reduced to two-thirds linear in reproduction.)

[*Jenkins & Mellor.*

CARBONYL IRON.

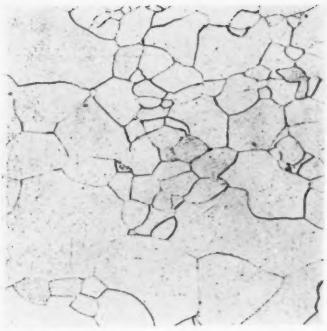

FIG. 107.—450° C. 19·9 tons per sq. in. Short-time tensile test. *C.* 253. Section $\frac{1}{8}$ in. from fracture. × 150.

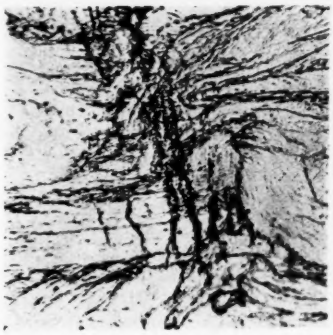

FIG. 108.—450° C. 19·9 tons per sq. in. Short-time tensile test. *C.* 253. Section close to fracture. × 500.

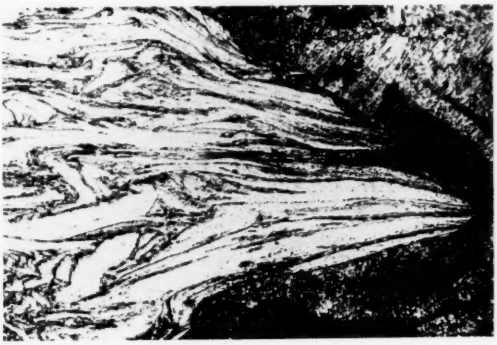

FIG. 109.—450° C. 5·6 tons per sq. in. Creep test. 3 hr. 15 min. *C.* 168. Section at iron-plated fracture. × 150.

FIG. 110.—550° C. 7·38 tons per sq. in. Short-time tensile test. *C.* 254. Section close to fracture. × 500.

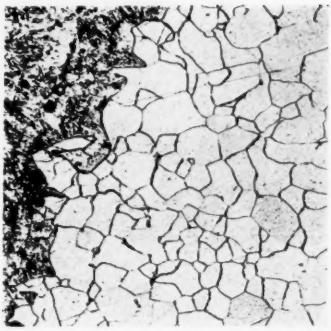

FIG. 111.—650° C. 3·84 tons per sq. in. Short-time tensile test. *C.* 257. Section at iron-plated fracture. × 150.

(Micrographs reduced to two-thirds linear in reproduction.)

[Jenkins & Mellor.

CARBONYL IRON.

FIG. 112.—650° C. 3·84 tons per sq. in. Short-time tensile test. *C.* 257. Section $\frac{3}{8}$ in. from fracture. × 500.

FIG. 113.—750° C. 0·5 ton per sq. in. Creep test. 1 day 2 hr. *C.* 255. Section at iron-plated fracture. × 150.

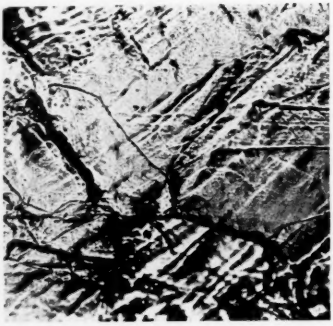

FIG. 114.—850° C. 0·9 ton per sq. in. Short-time tensile test. *C.* 274. Surface $1\frac{1}{2}$ in. from fracture. × 300.

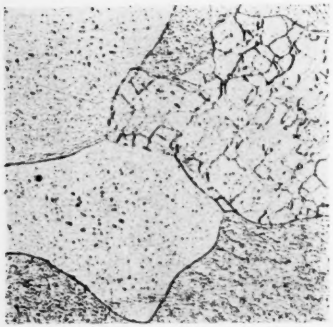

FIG. 115.—850° C. 0·9 ton per sq. in. Short-time tensile test. *C.* 274. Section $\frac{3}{8}$ in. from fracture. × 150.

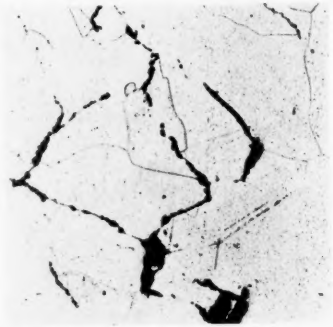

FIG. 116.—950° C. 0·35 ton per sq. in. Creep test. 19 hr. *C.* 220. Section $\frac{1}{4}$ in. from fracture. × 150.

(Micrographs reduced to two-thirds linear in reproduction.)

[*Jenkins & Mellor.*
[*To face p.* 225.

into rounded globules, which have little stiffening effect on the continuous ferrite phase.

The rates of spheroidisation of carbon steels at temperatures from 500°–700° C. have been discussed by Bailey and Roberts,[15] who suggest the following equation for the variation of the rate of spheroidisation with temperature :

$$t = A\,(e)^{\frac{33{,}000}{T}}$$

where t is the time in hours required to bring about a given degree of spheroidisation, T is the absolute temperature, and A is a constant influenced by the thermal history and mechanical condition of the material. In their tests on the effect of creep on the rate of spheroidisation, these authors employed rates of creep which resulted in a very small total strain, and they considered that the strain set up by working stresses had an almost negligible influence on the rate of spheroidisation (see their tests at 560° C. and 600° C.). The present work shows that in 24 hr. at 550° C. spheroidisation under the deformation produced by a load of 4 tons per sq. in. is quite appreciable and in marked contrast to the inappreciable amount produced in an unstressed specimen. At 450° C. and 550° C. in the absence of appreciable deformation spheroidisation is extremely slow, but becomes somewhat rapid in the vicinity of fracture in long-time creep specimens. It thus appears that the rate of diffusion of the carbide has been greatly increased by the deformation. The relative rates of recrystallisation and of diffusion of the carbide differ at various temperatures, and it is unlikely that spheroidisation at 450° C. will ultimately have the same effect on the properties of a material as that which is effected at 650° C.

The change in the lattice structure of iron ($\alpha \rightarrow \gamma$ change) gives rise to appreciable alteration in the strength of the material, in the microstructure, and in the mode of deformation. In the work of Rosenhain and Humfrey[2] the tests on an almost carbonless iron showed a marked fall in strength with increasing temperature to a minimum at about 840° C. The present work on Swedish iron, Armco ingot iron, and carbonyl iron confirms these results as regards the change of strength of these materials. Mild steel, however, has been found to be stronger at 850° C., a temperature which is above the A_3 critical point, than below it at 750° C.

The samples of Swedish iron and of Armco ingot iron contain larger and more numerous non-metallic inclusions than the mild steels. The carbonyl iron is apparently free from included matter. In no case does it appear that the mode of deformation and fracture has been directly influenced by individual pieces of non-metallic material. An indirect effect is undoubtedly

present in the case of Swedish iron which tends to split into longitudinal fibres parallel to the direction of the non-metallic inclusions. Intercrystalline cracking of Armco ingot iron is not connected with visible included matter, but may possibly be associated with the oxygen and nitrogen contents and a proportion of other constituents.

The effect of the process of manufacture of the various materials is revealed by differences in the mode of deformation in long-time creep tests. Mild steel shows little tendency to develop intercrystalline weakness or to split longitudinally into fibres; it thus appears that the presence of carbon during manufacture has kept the material free from some of the defects which arise from the presence of oxides. Swedish iron shows ease of recrystallisation, considerable ductility, but a marked tendency to form longitudinal fibres under the action of deformation. The mode of manufacture causes an unusual distribution of the impurities in Armco ingot iron, which is the least ductile of all the materials and is characterised by a tendency to develop intercrystalline cracking at high temperatures. Carbonyl iron shows no tendency to split longitudinally into fibres, but is otherwise very similar in behaviour to Swedish iron.

The authors are grateful to the late Dr. W. Rosenhain for suggestions in the initial stages of the work, and to Dr. C. H. Desch, F.R.S., Superintendent of the Metallurgy Department of the National Physical Laboratory, during the later stages. They also wish to recognise the assistance of Mr. E. A. Jenkinson, B.Sc., which has been of great value throughout the whole of the experimental work.

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DISCUSSION.

Mr. L. E. BENSON (Manchester) said that the paper was full of information and represented the beginning of a systematic study of a group of phenomena which were related and very complex. It had been obvious for some time that some such systematic study was due. He thought the authors were very much to be congratulated on the information they had presented, and that thanks were due to them and also to the Department of Scientific and Industrial Research. If he appeared to criticise the paper, he hoped the criticism would be accepted in the friendliest way.

In their discussion of results the authors referred to the intercrystalline type of cracking which was such a fascinating study, and they mentioned that "failure by intercrystalline cracking is found in materials subjected to internal strain accompanied by corrosive conditions." Evidently the season-cracking type of failure was referred to, but why *internal* strain? The season-cracking type of failure could occur with external just as with internal strain. If external strain were in mind, the analogy between that type of cracking and the type that might occur under creep conditions was more marked, though he very much doubted whether the analogy was well founded. Mr. Benson thought the authors did not really intend to suggest that it was well founded in the paper.

In the same paragraph of the paper, the authors remarked that on creep-testing specimens of alloy steels the crystals had been stiffened by alloy additions, but that this did not appear to strengthen the grain boundaries, the suggestion being that the crystalline material behaved something like a brick wall in which the bricks had been made hard and strong but the mortar remained weak. That was an idea which was a useful one as a picture, but again he felt it did not help greatly to understand what occurred in the intercrystalline failure of steels under creep conditions. It might perhaps represent what occurred in some measure, but the phenomenon which was being considered was probably more complex than the breaking down of a brick wall built with weak mortar.

The information given regarding the behaviour of Armco ingot iron, and comparing the breakdown of Armco ingot iron with the mode of failure of other irons, was most illuminating and suggestive. The authors said, in that connection, that the peculiar behaviour of the Armco iron was probably connected with the distribution of oxygen, nitrogen and other constituents, which led to a stiffening action on the ferrite. He suggested that that

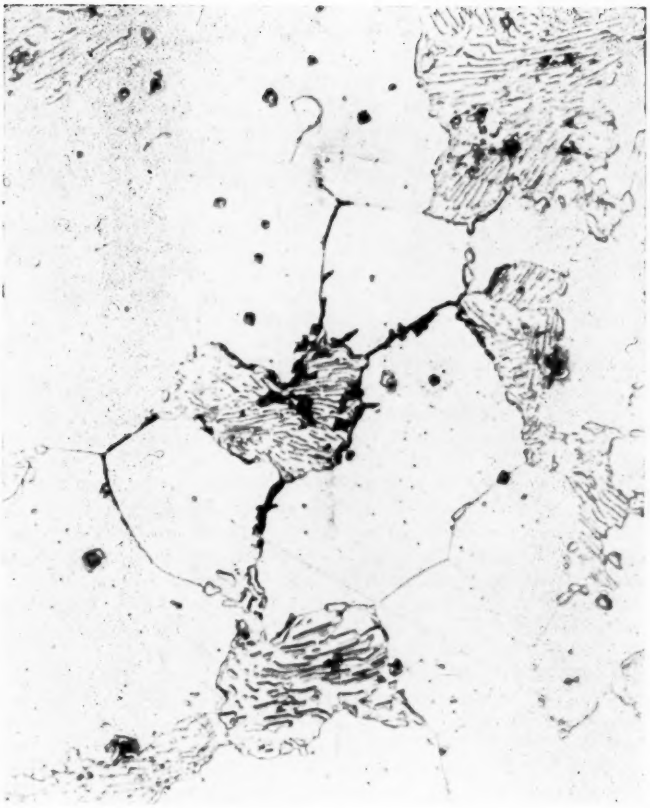


FIG. A.—Intercrystalline Graphitisation of Steel containing 0·5 per cent. carbon, 1 per cent. cobalt, graphitised by heating for 393 hr. at 625° C. × 800.
(*See* Benson's Discussion.)

[*Jenkins and Mellor, Discussion.*

[*To face p.* 228.

was a debatable statement. He thought that the cause or mechanism of this intercrystalline type of failure was simply not understood, and it was desirable to keep a perfectly open mind about it for the present.

His point was emphasised by the illustration in Fig. A (Plate XXXVIIIA.), which was of a sample of steel taken from some of the work on spheroidisation with which he had been connected. The particular sample in question showed a piece of steel which had been subjected to a certain temperature for a certain time, together of less intensity than might occur in the lifetime of a highly superheated steam turbine. The steel had quite obviously graphitised, and it was evident from that sample and from others that graphitisation of steel at superheated steam temperatures was a phenomenon which should be borne in mind, and was another possible mode of failure which apparently had not been considered by anyone so far. It was particularly interesting that the graphitisation, when it occurred under those conditions, might show a very definite trend along the crystal boundaries rather than within the crystals. The illustration revealed a definitely intercrystalline network of graphite. He was not suggesting that that was the explanation of intercrystalline cracking; he merely mentioned it to illustrate the point that other possible explanations of intercrystalline failure might exist, different in nature from that postulated by the authors.

The only other point to which he desired to refer was with regard to spheroidisation. The authors said (p. 224) "If spheroidisation is continued to completion, the iron carbide is aggregated into rounded globules, which have little stiffening effect on the continuous ferrite phase," the inference being that steel was stiffened in the sense of resistance to creep mainly by the lamellæ or globules of carbide. Just how far the stiffening was in fact due to the carbide was, he thought, an open question. He thought the behaviour of material under creep conditions was probably not related simply to the carbide—at least not to the carbide visible under the microscope. One could have a sample of steel with, say, an overheated structure in which the pearlite and the ferrite were very well defined in separate areas. Such a sample might have good creep resistance, whilst the same steel with a hardened and tempered structure where the carbide was apparently more evenly dispersed might have a comparatively poor creep resistance. He felt, therefore, that that inference was not quite justified.

DR. C. H. DESCH asked whether Mr. Benson could give the composition of the steel shown in Fig. A.

MR. BENSON: Yes. It contained about 0·5 per cent. of carbon and 1 per cent. of cobalt, he thought, but he would confirm that.

DR. DESCH suggested that it was very different from the soft, low-carbon materials without any cobalt and nickel.

MR. BENSON agreed, but added that he had found the same phenomenon in other steels as well, though not, so far, in plain carbon steels.

CORRESPONDENCE.

Mr. J. A. JONES (Sheffield) wrote that anyone who had had experience in the microscopical examination of strained or fractured test-pieces would admire the diligence shown by the authors, and they were to be congratulated on the quality of the large number of photomicrographs reproduced in the paper. The paper constituted a valuable contribution to the study of the prominent features of the changes which occurred in the structures of steels as a result of the application of stress within a wide range of temperature.

The authors' statements and observations were generally acceptable. There were, however, certain items concerned with the results of the mechanical tests which seemed to require amplifying.

It would be of considerable assistance, for purposes of comparison with results obtained in the laboratories of the companies with which the writer was associated, if the rate of straining in the Amsler machine could be stated more precisely than " a life of 2 to 3 min."

It was interesting to note the observation made on p. 187 that steels of higher carbon content were not superior in creep resistance to mild steel. Was this meant to be a general statement with regard to carbon content, applicable to any type of steel ? White, Clark and Wilson dealt very fully with this question in a paper presented to the American Society for Metals, entitled " Influence of Carbon Content on High-Temperature Properties of Steel " and seemed to arrive at the conclusion that an increase in the carbon content increased the " short-time strength " but that the creep properties were more dependent on the condition of the carbide than on the actual carbon content.

Turning to the results given in Table III. for steel *NGK*, it was extremely unfortunate that each one of the creep test-pieces tested at 450° C. fractured outside the gauge length. It was, therefore, important to know whether the authors' statement with regard to low ductility in this range was based on some other observations than low percentage elongation figures. It would appear from the figures for the reduction of area that the ductility of this steel compared favourably with that of the mild steel *M* tested at the same temperature.

It was extremely desirable that they should know with reasonable certainty whether certain mild steels were susceptible to loss of ductility when subjected to stress at such an important temperature as 450° C., while other mild steels were immune.

It would have added considerably to the value of this section of the paper if the suggested loss of ductility at 450° C. had been verified in steel *NGK* and an attempt made to discover the reason for an apparent difference in the behaviour of the two mild steels when tested at 450° C.

Previous investigators had noted the loss of ductility at temperatures approaching the Ac_3 change, while the increase in tensile strength above the A_3 point had also been established. This effect was very clearly indicated for a stainless steel and for a high-carbon chromium-molybdenum steel, in a paper entitled "Choice of Steels for Motor Valves" by Grard and Michel, *First Congress on Aerial Safety*, 1931. Similar results had been obtained for stainless steels in the writer's own laboratories. The well-known superiority of the austenitic steels from the point of view of strength at high temperatures was naturally associated with the feature of the increase in tensile strength obtained at temperatures above the transition to the austenitic condition.

It was interesting to note that in spite of a marked difference in strength between mild steel *NGK* and the Swedish iron at 850° C.—4·67 and 1·55 tons per sq. in. respectively in the Amsler machine—the tensile strength, in the austenitic condition, at 950° C. was identical for the two steels, while the value obtained at the same temperature for Armco iron was again approximately the same.

The statement that Armco iron was the only material examined which was really prone to intercrystalline cracking was particularly interesting. Did the authors infer that material which was liable to fracture in this manner was unsuitable for service at an elevated temperature? If such was the case, was it considered undesirable to introduce into steel for high-temperature service an element which had a stiffening effect on the ferrite alone and thus increased the liability to intercrystalline cracking?

Particularly interesting observations were made on p. 223 regarding the slip observed in austenitic steels of the 18/8 type. Similar results had been noted in the writer's own laboratory when determining the stress-time relationship in the Barr-Bardgett machines, by applying equal increments of strain at intervals of 1 min. and measuring the stress at the end of each interval. The diagram obtained by plotting the stress against time represented a stress-strain relationship for the material. In a test conducted at 500° C. it was found that after exceeding the elastic limit slipping occurred, and increased in amount as the strain increased. At certain stresses, slipping would take place

suddenly, the resulting stress being less than that reached by the preceding increment. A similar test on the same material carried out at 600° C. resulted in a smooth curve with no indication of slip. It was interesting to note that in the test at 500° C., the slipping was accompanied by a sound similar to that produced by the fracture of a test specimen. The apparatus was particularly insensitive to vibration effects so far as stress was concerned.

It would be interesting to know if the authors had observed this phenomenon to such a marked extent in other steels, and if there was a particular temperature at which the slipping ceased and their opinion of the significance of the phenomenon.

Mr. J. H. WHITELEY (Consett) wrote that he considered the authors had contributed a very instructive and valuable paper. The degree of experimental skill and patience exhibited was deserving of high praise, as also was the excellent discussion of the results obtained.

During the perusal of the paper a number of points had occurred to the writer, to some of which he wished to draw the authors' attention.

The first was the extraordinary difference between the tensile strengths of the two steels *NGK* and *M*, which certainly required explanation. It was, of course, a common experience in works practice that basic mild steel was somewhat softer than acid steel of similar analysis, but not to the extent here shown. There was nothing in the analyses given to which this difference could be assigned. The nitrogen content of steel *M* was, the writer thought, abnormal. In acid steel it seldom exceeded 0·005 per cent., as Dr. Swinden had shown in the Sixth Report of the Heterogeneity Committee. That figure was quite in accord with numerous estimations he (the writer) had made from time to time. The sulphur content, on the other hand, was low for acid bar steel. As regards the analysis in general, it was surprising that in such a thorough examination the arsenic and tin contents were not included.

He (Mr. Whiteley) found the authors' observations on the rapid spheroidisation or divorce of pearlite in stressed material particularly interesting, especially as they were in agreement with some work he did on the subject many years ago. In that investigation the steels were deformed at room temperature and then heated, whereas the deformation in the present work was at the temperature of heating. The effects, nevertheless, appeared to be the same in the two cases, and it could be concluded from the authors' results that heavy cold-rolling at a black heat in a hot-rolling mill could start a similar modification of the pearlite. There was still scope for research, however, on the influence of the degree of stress, temperature, time, and size of pearlite lamellæ on the speed

of spheroidisation. Were the authors quite certain that the marginal coalescence shown in Figs. 36 and 53 was not present in the steels before testing? Such effects were frequently to be observed in steels which had been cooled at a moderate rate from above Ar_3.

The formation of cavities in the stressed pieces at the higher temperatures was, he thought, remarkable, and it would be interesting to know whether these cavities could be detected in the polished specimens before etching. Did the authors observe any slow liberation of gases at these temperatures? Finally, the statement was made in the summary that "if spheroidisation is continued to completion, the iron carbide is aggregated into rounded globules, which have little stiffening effect on the continuous ferrite phase." Such a conclusion was, he thought, not apparent from the tables and graphs given. These data showed that at 15° C. the tensile strength of the steel *NGK* was 25·4 tons and the average strength of the Armco and Swedish irons was about 20 tons. Hence, it might be concluded that the carbon and manganese contents of this steel had increased its strength above that of ferrite by less than 30 per cent. With steel *M* the increase was about 40 per cent. Now at 550° C. ten-day creep test results show that the tensile strength of the two steels was over 50 per cent. greater than that of the two irons, and at 650° C. the percentage increase due to carbon and manganese was still as high as at room temperature. When the few short-time Amsler tests at these temperatures were examined similar figures were obtained, from which it would appear that the proportional increase in strength imparted by the carbon was not diminished at temperatures up to 650° C.

Mr. A. B. Winterbottom (Trondheim, Norway) wrote that the mass of valuable data recorded in the paper would help very materially towards a fuller understanding of the mechanism of creep deformation and ultimate failure. In his opinion, however, the temptation to draw too comprehensive conclusions from tests in which the various factors had been rightly limited and controlled, should be avoided. For instance, the behaviour of Armco iron, the effect of accidental air leakage in the case of mild steel *NGK* and the surface effects in general seemed very significant. Surface phenomena might indeed be determinative under normal service conditions. It would, therefore, be useful to have information concerning the deformation and structural changes in the same materials when tested in air and/or steam; or in the absence of precise data an opinion based on the authors' experience.

The reference to the use of X-rays in determining lattice strains suggested that such measurements made in the course of creep tests would enable the partition of the applied stress between

elastic and plastic strain to be continuously followed. Alternatively, if surface cracking and release of stress precluded such measurements the technique might still be of value as a sensitive means of detecting surface damage in creep and fatigue tests.

AUTHORS' REPLY.

The AUTHORS, in reply to Mr. Benson, considered that internal strain was produced in material by either internal stress or externally applied stress. They did not wish to suggest an analogy between season-cracking and the form of cracking found in Armco ingot iron. In the paper, two instances of intercrystalline weakness were pointed out from the many in which this form of cracking was prevalent.

The analogy to bricks and mortar had certain obvious objections, but the term had sometimes been used for convenience. The "mortar" was not necessarily weak, but the authors did suggest that the "mortar" was not strengthened by alloying elements to the same extent as was the material within the grains.

The creep characteristics of Armco ingot iron appeared to be similar to those found in alloy steels, and, therefore, it was thought that these characteristics were due to the proportions of oxygen or other constituents which were present in greater amounts than those found in the other steels and irons. The average results of the oxygen analyses of the four materials used in this work were as followed: Armco ingot iron, 0·18 per cent.; Swedish iron, 0·34 per cent.; mild steel, 0·014 per cent.; carbonyl iron, 0·0019 per cent. Although Swedish iron had the highest oxygen content, it was known that much of that oxygen was present in the form of large slag streaks, whereas in Armco ingot iron the oxygen content was fairly evenly distributed, both in regard to slag particles and the high proportion of oxygen held in solid solution. The oxygen content of Armco iron had two effects on the material. It reduced the ductility at high temperatures and raised the recrystallisation temperature.

The photomicrograph of a sample of cobalt steel was very interesting. It appeared unusual that graphite should be formed in a pearlite grain before the lamellar carbide had spheroidised.

It did not seem likely that graphitisation could be taken to be an explanation of intercrystalline failure in Armco ingot iron, because the carbon contents of the two samples tested were only 0·016 and 0·022 per cent., respectively, and it was not to be expected that graphitisation would take place in irons of such

low carbon contents. It had been found that graphitisation took place in some higher carbon steels—0·6 per cent. of carbon and above—during creep tests at temperatures as low as 550° C.

Mr. Jones asked for further information on the rate of straining of short-time tensile test-pieces. No record of the actual rate of straining was taken, but an approximate value for the rate could be obtained from the elongations given in the tables and the duration of the various tests which were plotted logarithmically in Figs. 5, 6, 8, 10, 11, 13, 14, and 16.

The statement on p. 187, that steels of a higher carbon content were not superior in creep resistance to mild steel, referred to carbon steels only and to temperatures of 550° C. and upwards. At temperatures up to 450° C., particularly at a high rate of strain, carbon did increase the creep strength of the material. The authors were in agreement with the conclusion of White, Clark, and Wilson quoted by Mr. Jones. The creep properties were more dependent on the condition of the carbide and the relation of the ferrite within the pearlite to the free ferrite in the neighbouring grains than on the actual carbon content.

No confirmation had been obtained of the low ductility of the mild steel *NGK* in creep tests at 450° C., but evidence had been obtained of a decreased ductility in notched-bar impact tests at room temperature of a mild steel which had commenced to spheroidise by treatment at 650° C. In the present work mild steel *M* did not show decreased ductility at 450° C.

The increase in tensile strength obtained at temperatures above the A_3 point was found by Rosenhain and Humfrey,[1] who concluded " that γ-iron, as found in approximately pure iron at high temperatures, possesses the characteristic structure and properties found in the ' γ-iron ' alloy steels."

In comparing the strengths of mild steel and Swedish iron at 850° C. it was interesting to note that mild steel *NGK* had passed the upper critical point, and in the γ condition had an ultimate tensile strength of 4·67 tons per sq. in. Swedish iron was still in the α condition at 850° C. and relatively weak, having an ultimate tensile strength of 1·55 tons per sq. in.

The suitability of a material for service at a temperature range in which intercrystalline cracking occurred was likely to be a subject of considerable controversy in the future. It had been found that in such cases there was a definite relationship between the stress and time for this type of failure to occur. The service stresses likely to be adopted were low in comparison with the stresses required to produce failure in test specimens at 450°–550° C., and failure would occur only after a very considerable interval of time which was generally much in excess of

[1]Rosenhain and Humfrey, *Proceedings of the Royal Society*, 1909, A. vol. 83, pp. 200–203.

the life of the part. At the moment the authors did not know of any alloying element which did not stiffen the material within the grains to a greater extent than the material in the vicinity of the grain boundary.

The authors were interested to note the suddenness of the slip which occurred in the vicinity of the elastic limit in tests in the Barr-Bardgett machine. The phenomenon was most noticeable in 18/8 chromium-nickel steel, but it had not been observed in low-carbon steels and irons.

The authors had no further remarks to add regarding the effect of carbon on the strength of the two steels *M* and *NGK*. As pointed out by Mr. Whiteley, the high nitrogen content of steel *M* might account for the difference.

The marginal coalescence observed in Figs. 36 and 53 was not present in the original material as normalised in the form of 1-in. diam. bars. Presumably Mr. Whiteley referred to its occurrence, possibly by slight spheroidisation, during the cooling of fairly large masses.

The cavities formed in test-pieces stressed at the higher temperatures had been observed on polished sections in the unetched condition. Slow liberation of gas from the test-pieces had, in general, been noted, and at low temperatures, say, 400° C., this was liable to cause slight oxidation on the surface of test-pieces. No definite conclusions could be drawn from the present work regarding the strength of spheroidised material, but other work in progress at the National Physical Laboratory during the last four or five years showed that the rate of creep of spheroidised material was very much higher than that of normalised material for similar testing conditions.

Mr. Whiteley's analysis of creep and tensile test results did not disprove the authors' conclusion that spheroidised carbide had little stiffening effect on the continuous ferrite phase. It should be noted that the manganese content of the mild steels was over 0·60 per cent., whilst that of the Swedish, Armco and carbonyl irons was below 0·035 per cent. The high manganese content of the mild steels probably led to a stiffening of the ferrite over a considerable range of temperature.

The authors would like to thank Mr. Winterbottom for his interesting suggestions. At present they were not in a position to indicate the degree of alteration of structure of material tested in air and in steam, but tests in high-pressure steam were now being carried o t which should lead to information at a later date.

MORPHOLOGY OF THE INCLUSIONS IN SIDERURGICAL PRODUCTS.[1]

By Professor A. M. PORTEVIN and RENÉ CASTRO (Paris).

Abstract.

In their introduction (Part I.) the authors draw attention to the value of the study of inclusions in tracing the previous history of the material in which such inclusions occur. They then go on to discuss microscope technique and chemical analysis, especially the information afforded by the latter, and in connection w th which they refer to the residue methods commonly employed. Then follows an indication of the extent to which the chemical and microscopic examination of inclusions may be made to reveal the history of the metal in the molten and plastic stages.

In Part II. the authors describe, in detail, the morphological features of a number of inclusions found in samples of iron and of plain carbon steels. The materials investigated contained manganese, silicon and aluminium, either separately or together; in each case the type of inclusion exhibited when oxygen, or sulphur, or oxygen and sulphur together are present is discussed; the characteristic appearance of a large number of such inclusions when examined under the microscope is dealt with, and suggestions are made as to their mode of origin.

Part I.—Introduction.

The problem of the non-metallic inclusions in metallurgical products is one that has always engaged the attention of metallurgists, but particularly so during the last few years. Hitherto it was thought that, in certain cases, these inclusions might be connected in some way with the quality of the metal. That is why, for more than a hundred years, investigators have worked on the problem in the endeavour to ascertain these relations and to determine, accurately, the constituents of non-metallic inclusions, especially sulphur and oxygen.[2]

There is, however, another equally important factor connected with non-metallic inclusions, and to this less attention has been given. The authors refer to their connection with the history of the metal. The inclusions visible in a sample of metal are very often the manifestation of one or more phases of the process of manufacture, which suggests that they may be regarded as

[1] Received June 12, 1935.

[2] Castro and Portevin, *Revue de Métallurgie, Mémoires*, 1932, vol. 29, pp. 414–421, 449–469, 492–506, 553–564.

internal slags.[1] In certain cases, an examination of the inclusions in a metal, coupled with a knowledge of the complete chemical analysis and a microscopic examination of the elements of its structure, may enable the worker partially to retrace its previous history. To enable this to be done, however, it is necessary to trace the previous history of a number of particular cases, and observe the types of inclusions which most frequently accompany a particular method of manufacture.

The present authors do not claim to break new ground or to exhaust such a vast subject as the present one. Attempts have already been made, in a limited number of cases, to outline the relation between non-metallic inclusions and the previous treatment of the metal; and, on the other hand, accounts—more or less complete—of the morphology of inclusions have been published. Frequently, however, the information afforded to metallurgists by such researches is incomplete, for several reasons: (1) Very often, inclusions have only been observed in "synthetic" castings, obtained by adding, to a small quantity of remelted metal, additions in bulk of substances which give rise to non-metallic inclusions in the melt. Although this method may be interesting and provide information on certain special points, it often leads the observer to study inclusions which differ appreciably from those actually found in practice. (2) Descriptions have very often been given of the aspects of inclusions in samples of refined metals whose exact history is unknown. In this case, the influence of the thermal and mechanical history of the metal is added to its chemical history, and this greatly increases the number of variable factors and makes them difficult to interpret. (3) Unfortunately, the published photographs have often been bad, or at least incomplete, in very many interesting cases, due either to unsatisfactory methods of reproducing the originals, or to the original prints themselves being unsatisfactory, because microphotographic technique is generally different and more delicate when dealing with inclusions than it is when reproducing the aspects of metallic constituents.

In the present paper, the authors have endeavoured to show the reader a large number of characteristic types of inclusions, classified according to their probable nature, and they have tried as far as possible to show their probable mode of origin.

In a previous paper (Castro and Portevin, *loc. cit.*), the authors briefly outlined the numerous methods of investigation necessary for a knowledge of inclusions. None of these methods can be regarded as really perfect at present, and it has been possible to utilise only a few of them successfully. These will now be briefly summarised.

[1] Portevin and Perrin, *Journal of the Iron and Steel Institute*, 1933, No. I. pp. 153–187.

(1) *Microscopic Examination.*

A.—According to several writers, the polishing of metal samples for the examination of inclusions calls for a very special technique —dry polishing—which is necessary to prevent portions of the structure from being torn away or striæ forming due to the unequal hardness of the constituents that have to be polished. As a matter of fact, ordinary polishing (coarse-grinding on the grinding wheel or with emery paper, finish-polishing with two grades of alumina) is just as suitable, provided (1) only used emery paper is employed ; (2) the specimens are not pressed too hard on the abrasive surfaces ; and (3) the specimens are finished off for a sufficient length of time with almost dry felt.

The appearance of the photomicrographs accompanying this paper shows that this method is sufficient in the majority of cases.

B.—Ordinary micrographic examination (and photography) at powers up to about 2,000 diam. was carried out in reflected light, using vertical glass plate illumination. The following methods were also utilised :

(*a*) Slightly oblique illumination (prism), which enabled interesting information to be gleaned as to the relative relief of the several constituents of the inclusions.

(*b*) Conical illumination by parabolic mirror (dark-ground illumination). This method of observation is very valuable because it permits of :

(1) Revealing very fine inclusions which are practically invisible by direct observation (principle of the ultramicroscope).

(2) Bringing out the degree of transparency and the true colour of the inclusions.

(3) Examining the often complex internal structure of transparent or translucent inclusions—a structure which is sometimes invisible in reflected light.

(*c*) Illumination by a narrow, centred pencil of light. Where coarse inclusions are being examined, it enables certain internal constituents to be distinguished which are otherwise not easily visible on account of reflection, diffraction and diffusion inside the translucent non-metallic substance.

(*d*) A comparison between examinations made with dry high-power objectives and with immersion objectives affords a qualitative guide as to the reflectivities of the non-metallic compounds. Moreover, as regards certain semi-transparent inclusions of large dimensions (glasses crystallised to a greater or lesser extent), the immersion objective enables the observer to have between the actual interior of the inclusion and the objective, a medium of steady refractive index which enables

the internal elements to be examined simply by varying the focus, without being troubled by reflections from the polished surface.

(*e*) Illumination in polarised light by reflection, between crossed nicol prisms. This method of illumination, which is now being adopted to a greater extent, [1, 2, 3, 4] supplies except for certain details, the same information as that afforded by dark-ground illumination with a conical mirror. It also enables the isotropy or the anisotropy of the constituents studied to be defined, even where they are opaque. At low powers, illumination by polarised light is not so bright as conical illumination ; but the reverse is true at high powers, as the short focal distance of the high-power objectives does not, in this case, enable the specimen to be illuminated from the outside. Illumination by polarised light (Hoyt and Scheil, *loc. cit.*) also eliminates the troublesome reflections set up by the pits and the polishing marks, as well as the considerable diffraction and chromatic aberration at high powers in oblique light. However, the authors used dark-ground illumination in the majority of cases, reserving examinations in polarised light for special or delicate cases (double refraction, examinations with immersion objectives).

C.—One other additional item of information as to the nature of the inclusions met with is afford d by etching the constituents of the inclusions ; but this only applies to a few cases, and only if the observer is very prudent and circumspect in interpreting the results.[5] This method was applied by the classic technique outlined by Wohrman[6] and by Campbell and Comstock.[7]

(2) *Chemical Analysis.*

In addition to the total analysis of the metal, which forms a preliminary guide as to the nature of the inclusions which it may be expected to contain (although such an analysis is incomplete, since certain elements such as oxygen, aluminium, calcium, &c., which act on the inclusions are rarely, if ever, determined) the ideal thing would be to have the chemical analysis of the actual inclusions. This, unfortunately, is impossible in the majority of cases because :

[1] Baeyertz, *Transactions of the American Society for Metals*, 1934, vol. 22 pp. 625–634.

[2] Hoyt and Scheil, *Metals Technology, Technical Publication No.* 567, Sept., 1934.

[3] Dayton, *Metals Technology, Technical Publication No.* 593, Jan., 1935.

[4] Urban and Chipman, *Transactions of the American Society for Metals*, 1935, vol. 23, Mar., pp. 93–112.

[5] Grant, *Transactions of the American Society for Steel Treating*, 1931, vol. 19, pp. 165–181.

[6] Wohrman, *Transactions of the American Society for Steel Treating*, 1928, vol. 14, pp. 81, 255, 385, 539.

[7] Campbell and Comstock, *American Society for Testing Materials*, 1923, vol. 23, Part I. p. 521.

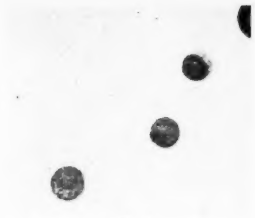


FIG. 1.

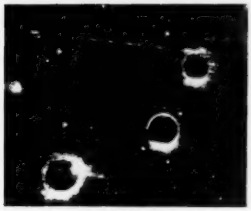


FIG. 2.

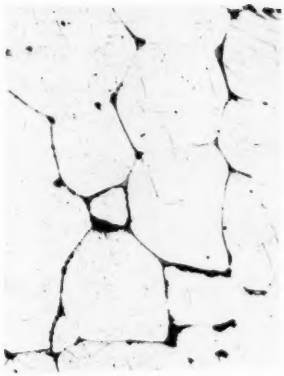
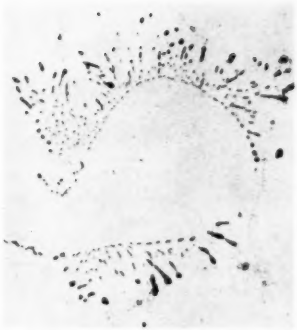


FIG. 3.



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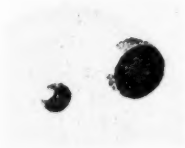


FIG. 5.

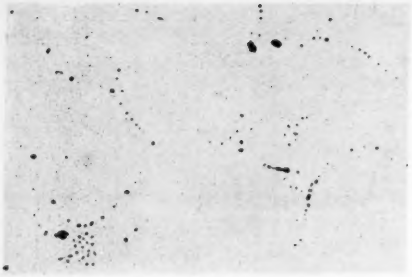
FIG. 6.

FIG. 7.

FIG. 8.

For descriptions see end of paper.

[*Portevin & Castro.*
[*To face p.* 240.



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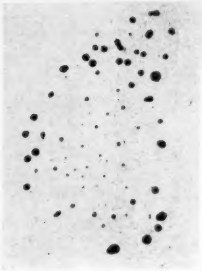
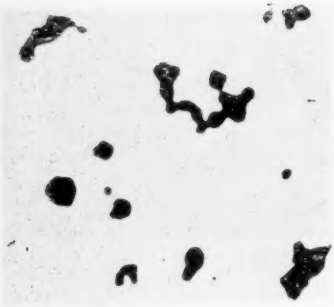


FIG. 10



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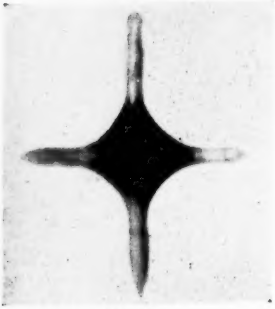

FIG. 13

FIG. 12.

FIG. 14.

FIG. 15.

FIG. 16.

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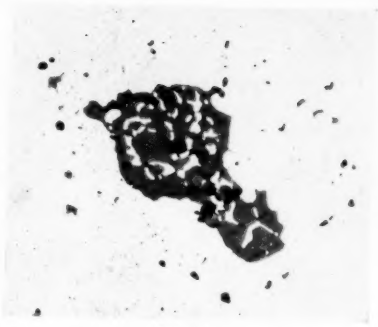
[*Portevin & Castro*



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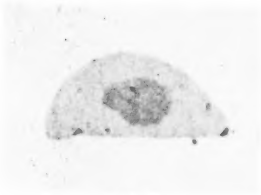

FIG. 19.

FIG. 21.

FIG. 20.

FIG. 22.

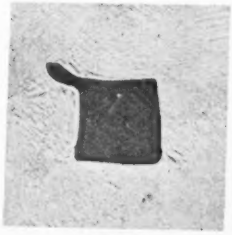

FIG. 23.

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[*Portevin & Castro*

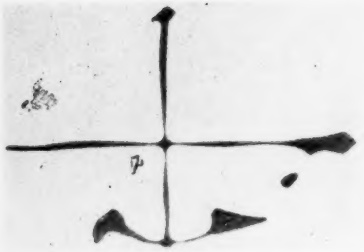
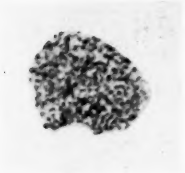


FIG. 24.



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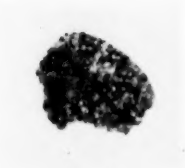

FIG. 26.

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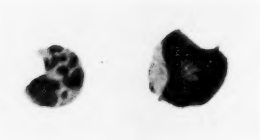

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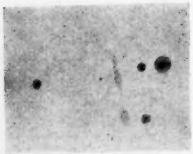
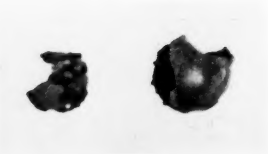


FIG. 30



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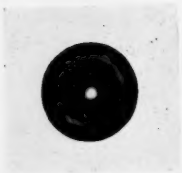

FIG. 31.

FIG. 32.

FIG. 33.

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[*Portevin & Castro.*

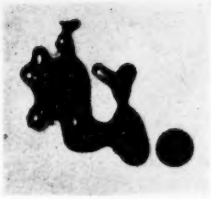
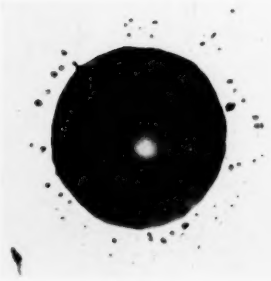


FIG. 34.

FIG. 35.

FIG. 36.



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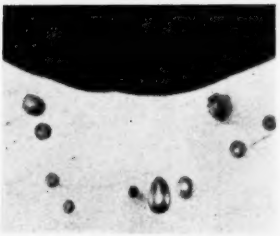

FIG. 38.

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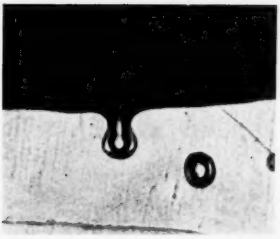


FIG. 40.

For descriptions see end of paper.

[*Portevin & Castro.*

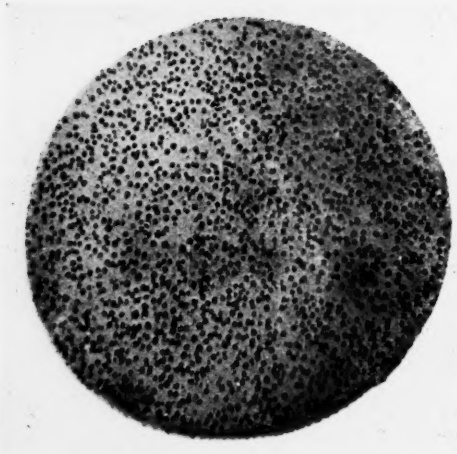


FIG. 41.

FIG. 42.

FIG. 43.

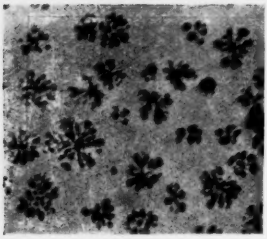


FIG 44.

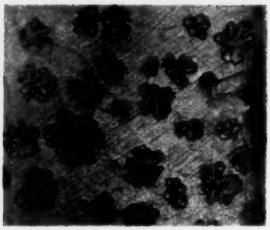
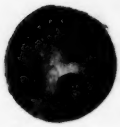


FIG. 45.

For descriptions see end of paper.

[*Portevin & Castro.*



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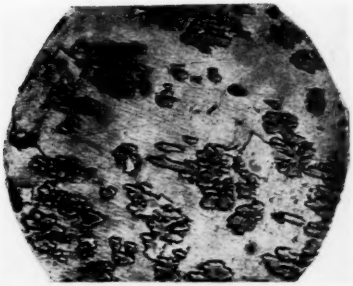

FIG. 47.

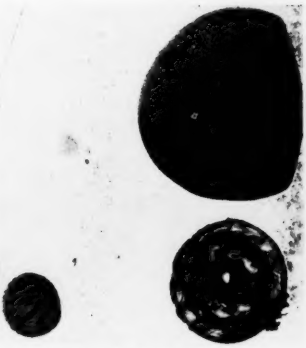

FIG. 48.

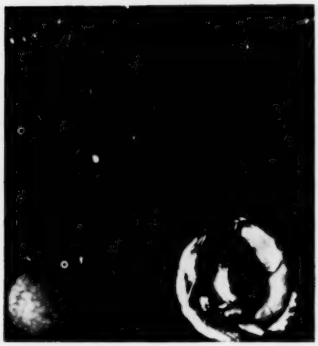

FIG. 49.

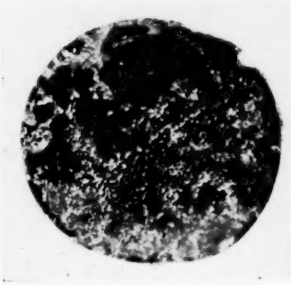

FIG. 50.

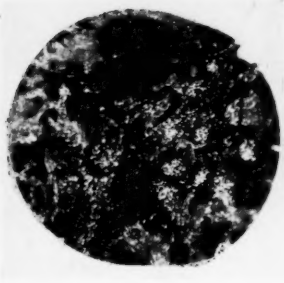

FIG. 51.

For descriptions see end of paper.
(Micrographs reduced to three-quarters linear in reproduction).

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FIG. 52.

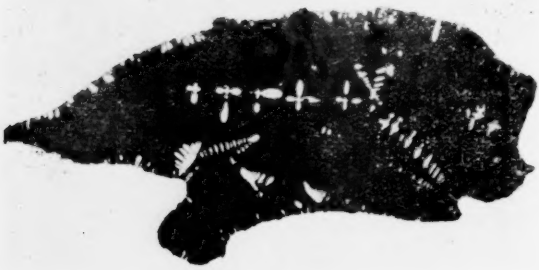

FIG. 53.

FIG. 54.

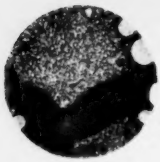

FIG. 55.

FIG. 56.

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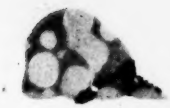

FIG. 57.

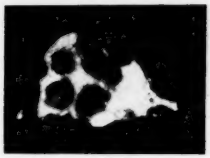

FIG. 58.

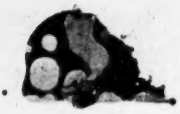

FIG. 59.

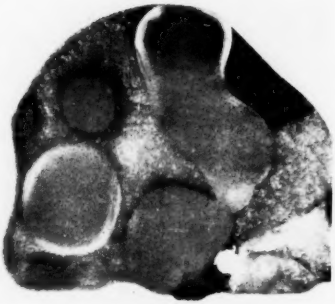

FIG. 60.

FIG. 62.

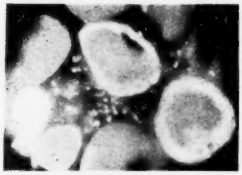

FIG. 61.

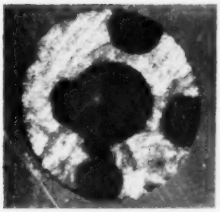

FIG. 63.

FIG. 64.

FIG. 65.

For descriptions see end of paper.

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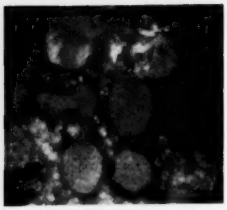

FIG. 66.

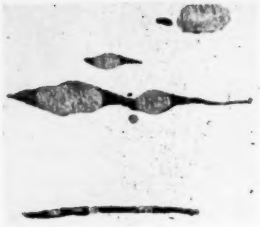

FIG. 67.

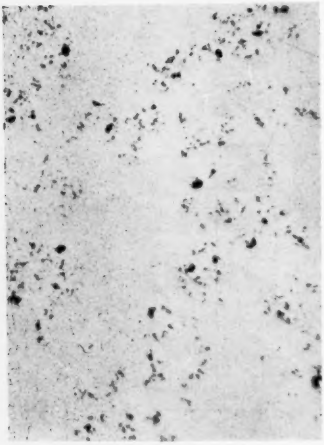

FIG. 68.

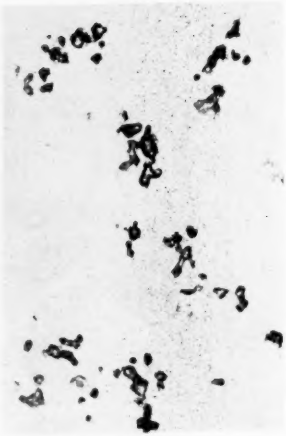

FIG. 69.

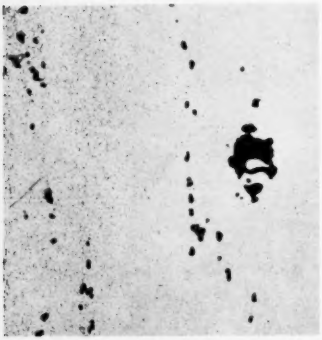

FIG. 70.

FIG. 71.

For descriptions see end of paper.

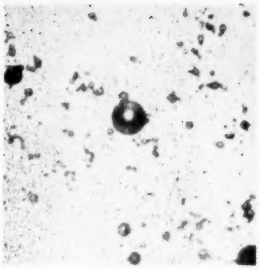

FIG. 72.

FIG. 73.

FIG. 74.

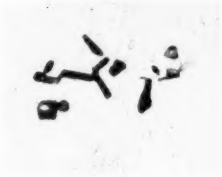

FIG. 75.

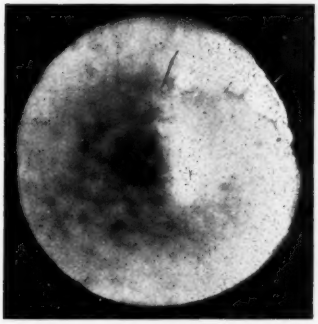

FIG. 77.

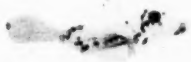

FIG. 76.

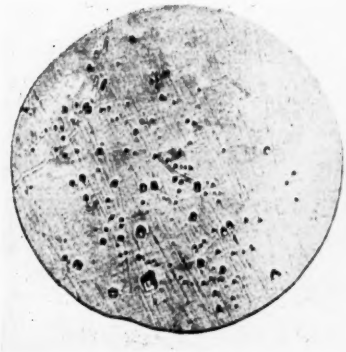

FIG. 78.

FIG. 79.

For descriptions see end of paper.

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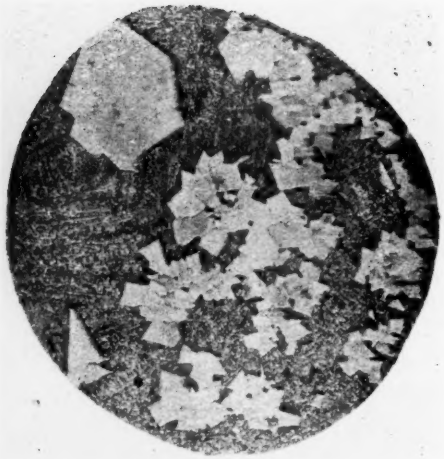

FIG. 80.

FIG. 81.

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FIG. 82.

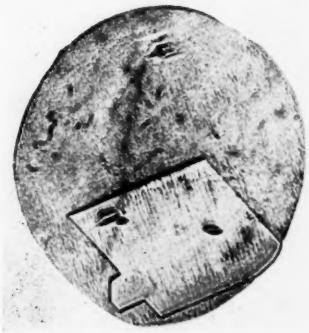

FIG. 83.

FIG. 84.

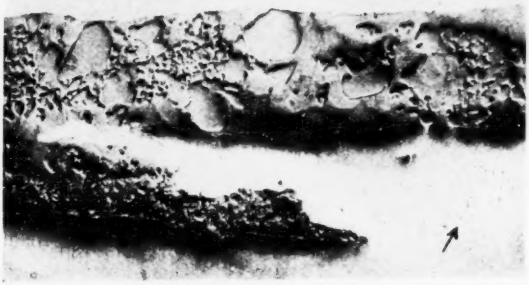

FIG. 85.

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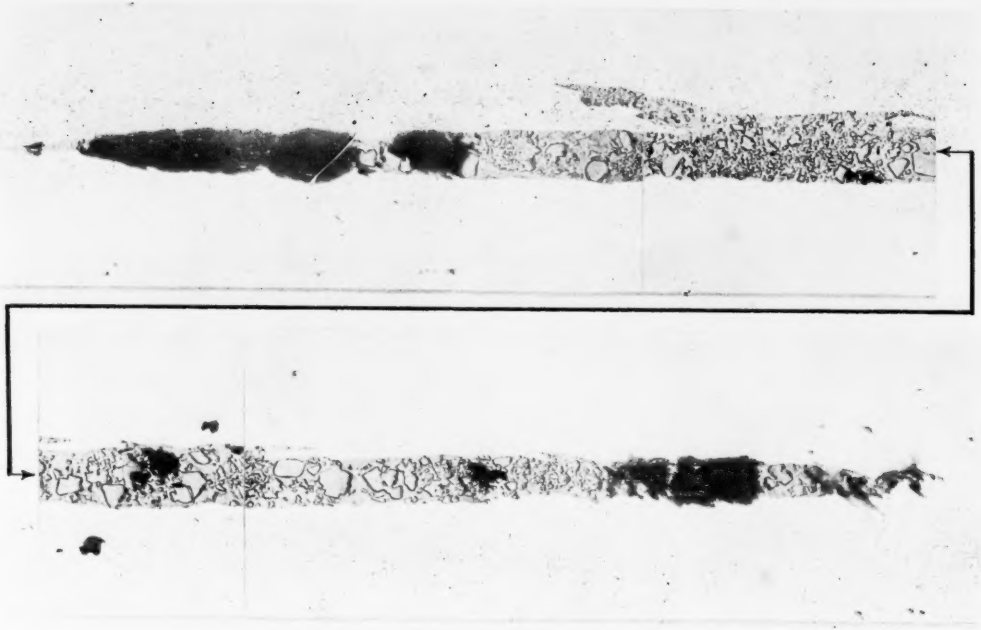

FIG. 86.

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FIG. 87.

FIG. 88.

FIG. 89

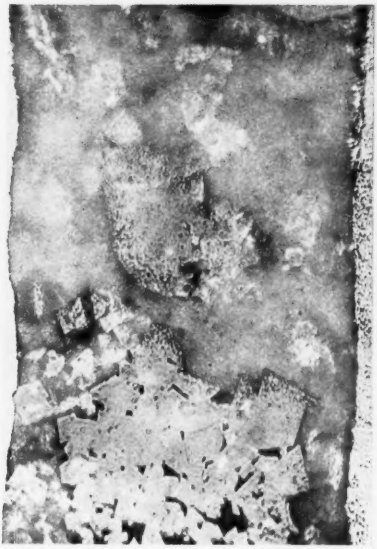

FIG. 90.

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FIG. 91.

FIG. 92.

FIG. 93.

FIG. 94.

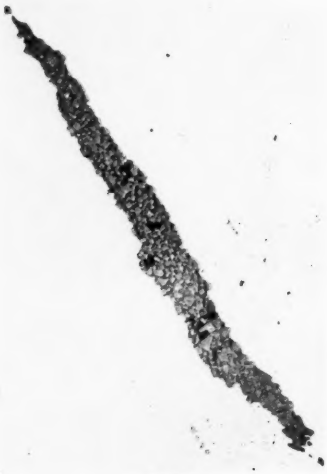

FIG. 95.

FIG. 96.

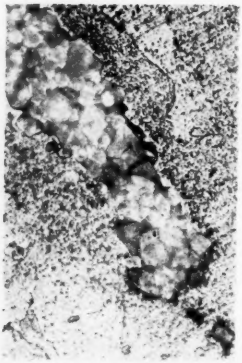

FIG. 97.

For descriptions see end of paper.

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FIG. 98.

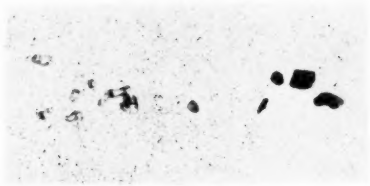

FIG. 99.

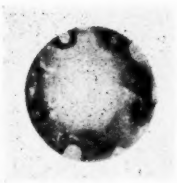

FIG. 100.

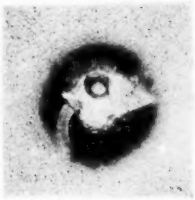

FIG. 101.

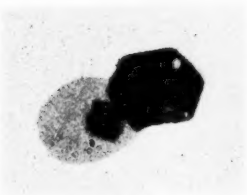

FIG. 102.

FIG. 103.

For descriptions see end of paper.

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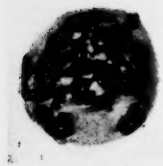

FIG. 104.

FIG. 105.

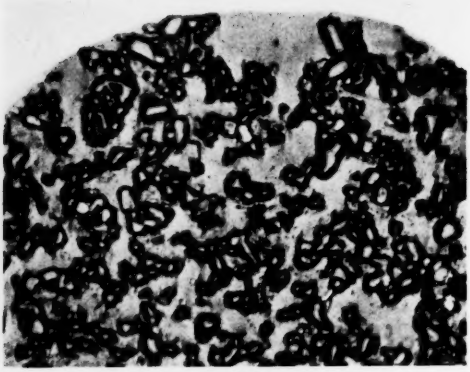

FIG. 106.

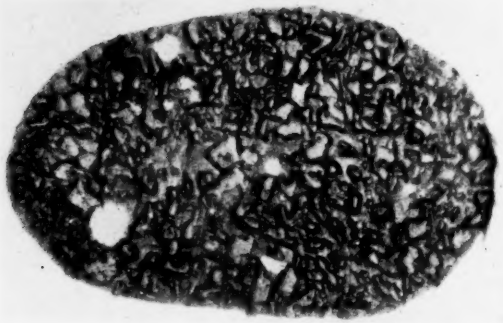

FIG. 107.

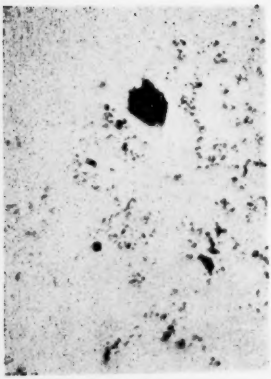

FIG. 108.

FIG. 109.

FIG. 110.

For descriptions see end of paper.

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[*To face p.* 241

(1) The influence of the chemical methods of separating the inclusions—what are called "residue methods"—admit of such analysis in a few cases only, which are limited (*a*) by the nature of the metal (refractoriness, the presence of structural elements contaminating the reacting residues), and (*b*) by the nature of the inclusions (constituents more or less affected by the reagent used).

(2) Even with a knowledge of one or more methods which enable the inclusions in steel to be isolated and analysed in every case, it often happens that the inclusions are of several types, or very complex. In that case, the analysis only gives the general composition of the inclusions in the test sample, and not the composition of each of the structural elements. The information supplied by analysing the residue is therefore incomplete, and similar to that furnished by the general analysis of a natural rock containing several constituents, for the purpose of studying it.

The authors have used the residue analysis method in only a limited number of cases where the following conditions were combined: (1) The presence of a single type of inclusion in the metal; and (2) where the inclusions were impervious to the usual isolating reagents such as HCl, Cl_2, FeI_2 (Castro and Portevin, *loc. cit.*). They have not yet used the electrolytic method[1] in cases where the metal was free from refractory structural elements (carbides of chromium, &c.).

The method of analysis used is the microchemical method elaborated by Treje and Alber.[2] This has proved very satisfactory, especially for determining alumina, as the ordinary methods are inaccurate in this case.

In every case the authors determined the total oxygen by the carbon reduction method *in vacuo* (which has now attained a high degree of accuracy[3]), in order to ensure that the total oxygen determined was as near as possible to the oxygen combined with the metallic elements of the residue; and the results were not included unless they agreed in this way.

(3) *Knowledge of the History of the Metal.*

(*a*) *Chemical History.*—By this is meant the changes in the chemical composition of the metal in the course of manufacture. Whenever it can be done, it is obviously useful to connect the variation in the appearance of the inclusions observed in the samples of metal taken whilst steel is being manufactured, with

[1] Treje and Benedicks, *Journal of the Iron and Steel Institute*, 1933, No. II. pp. 205–236.

[2] Treje and Alber, *Jernkontorets Annaler*, 1933, vol. 117, pp. 457–473.

[3] Practical improvements recently made in the technique and apparatus will shortly be published by the present authors.

the variations in chemical composition caused by the additions made, or by other metallurgical operations. As a matter of fact, this knowledge may supply information that the chemical analysis is unable to do (quantity, method and period of adding deoxidants, for instance). Again, the identification of certain exogenous inclusions met with, and their study, call for information regarding the slags and refractories that have been in contact with the metal.

(*b*) *Mechanical and Thermal History.*—The deformations which the metal undergoes in the solid or plastic state have a repercussion on the form of the inclusions. In this connection, the authors would like to call attention to the fact that a study of the genesis and structure of the inclusions can be made more easily and advantageously on metal as cast than it can on hot-worked metal. As a matter of fact, hot-working deforms the inclusions or breaks them up and makes their original appearance or their distribution unrecognisable. The heat treatment associated with hot-working may also cause the modification of certain structural elements. Whenever possible, therefore, it is advisable to examine the inclusions in metal as cast, and preferably in fairly large ingots, because, in that case, the inclusions are large (coalescence, &c.) and can be studied more easily. From this point of view, steelworks laboratories are better situated than university, technical college or private laboratories, as it often happens that the latter have only a fragment of metal, often hot-worked, to go upon and do not know its true history. Moreover, it is much easier to deal with special cases, which are often very instructive.

Once the investigation has been made on metal as cast, the same metal can be rolled or forged, and the following information derived from a fresh examination :

(1) The relation between the appearance before and after deformation. It should be noted that it is the appearance after deformation which is most often observed by the user.

(2) Conclusions regarding certain physical properties of the inclusions, such as fragility or plasticity in terms of temperature of deformation, method of distribution, transformation of certain elements of the structure, &c.

All this experimental data will then afford :

(1) A knowledge, purely morphological and as complete as possible, of the types of inclusions met with.

(2) Information regarding their probable nature. This information will be more or less complete according to the particular case dealt with.

The methods of investigation employed enable the worker to say with greater certainty whether a particular inclusion contains silica, alumina or oxide of chromium, rather than to make state-

ments regarding the precise mineralogical character (fayalite, mullite, sillimanite, olivine, &c.) of the constituent observed. Any statements that may be made under this head are very often based mainly on hypothesis and are so difficult to verify—if they can be verified at all—that they cannot be admitted unreservedly, and caution is always necessary in making affirmations on this point.

(3) Information regarding their mode of formation, which, in certain cases, enables the observer to go back to certain points in the history of the metal, where it is unknown.

Since the classic researches of Benedicks and Löfquist,[1] it is usual to distinguish between endogenous (or native) inclusions ("native slag grains"), formed within the metal during its manufacture, &c., and exogenous (or allogenous) inclusions ("foreign slag grains"), arising either from the refractory materials or the slag brought into the metal, or from a mixture of the products of deoxidation with these substances. The study of the exogenous inclusions is a very delicate and complicated matter, due to the enormous variety of their conditions of formation. For the moment, only the endogenous inclusions will be considered. As is known, it is chiefly oxygen and sulphur which are the main elements producing non-metallic inclusions in iron and iron alloys. Nitrogen and phosphorus form, with the metals, compounds whose characteristics, pretty much like those of the carbides, more nearly approach the characteristics of metallic substances and may therefore be regarded as true structural elements of the alloy. Certain special nitrides, however, closely approach the non-metallic inclusions in appearance and properties, and these will receive mention at the end of this paper.

The elements which promote inclusions are therefore mainly the metalloids oxygen and sulphur. The supporting elements with which sulphur and oxygen are combined to form non-metallic inclusions are iron and its various alloying or concomitant elements. Morphologically, the following distinction may be made:

A.—The inclusions in iron, either with or without its usual concomitant elements, *i.e.*, manganese, silicon and aluminium, separately or simultaneously.

B.—The inclusions in iron containing, besides the foregoing elements, special alloying elements (chromium, vanadium, titanium, &c.).

Part II.—Iron, Steels, and Alloys Without Special Elements.

The non-metallic inclusions set up by the presence in steel, unalloyed or alloyed, of varying quantities of manganese,

[1] Benedicks and Löfquist: "Non-Metallic Inclusions in Iron and Steel." London. Chapman and Hall, 1930.

TABLE I.—*Analyses of the Samples Examined.*

No.	Analyses.						Remarks.
	C. %	Si. %	Mn. %	S. %	P. %	O. %	
1	0·080	0·010	0·080	0·016	..	0·070	4 kg. ingot as cast. Taken from a 15-ton electric furnace melt in process of manufacture.
2	0·250	0·010	0·100	0·120	..	..	Electrolytic iron, melted in a 5 kg. H.-F. furnace, carburised and sulphurised.
3	0·035	0·010	0·130	0·026	..	0·110	4 kg. ingot as cast, taken from a 15-ton electric furnace melt during the course of a "heat."
4	0·235	0·010	0·100	0·010	..	0·011	Same remark as for No. 3.
5	0·035	0·010	0·280	..	..	0·080	Sample of basic Bessemer steel taken from the converter at the end of the after-blow period. Forged.
6	0·120	0·010	0·310	0·032	..	0·060	Same remark as for No. 3.
7	0·055	0·020	0·800	0·015	0·010	0·030	Same remark as for No. 3.
8	0·155	0·010	1·20	0·020	..	..	500 kg. ingot of electric steel. Rolled.
9	0·185	0·010	1·09	0·025	..	..	500 kg. electric steel ingot, as cast; 0·005 per cent. of aluminium added to the melt.
10	3·5	2·45	1·80	0·045	0·050	0·007	Hematite pig-iron.
11	1·1	0·10	1·15	0·040	..	..	Centre of crop-end of ingot No. 9.
12	0·030	0·495	Traces	..	..	0·065	Electrolytic iron melted in a 5 kg. H.-F. furnace, oxidised, silicon killed. As cast.
13	0·035	0·115	0·100	..	..	0·050	Same remark as for ingot No. 12. Metal highly superheated before casting. As cast.
14	0·040	0·710	0·615	..	..	0·054	Same remark as for ingot No. 12. Addition of silico-manganese. Forged.
15	0·120	0·350	0·390	0·022	0·010	0·009	500 kg. electric steel ingot, as cast.
16	0·090	0·215	0·485	0·017	0·010	0·015	30 kg. electric steel ingot, as cast.
17	0·250	0·120	0·550	0·010	0·019	..	500 kg. electric steel ingot, as cast.
18	0·085	0·210	0·495	0·017	0·010	..	500 kg. ingot. Same melt as No. 16.
19		Same	Ingot	as No.	18.		After rolling.

TABLE I.—(*continued*).

No.	Analyses.						Remarks.
	C. %	Si. %	Mn. %	S. %	P. %	O. %	
20	0·100	0·060	0·040	0·170	0·006	..	Gallo-Roman wrought iron.
21	0·120	0·020	0·450	0·058	0·035	0·021	Rimming open-hearth steel. Two-ton ingot, as cast.
22	0·120	0·010	0·420	0·031	0·031	0·032	Same ingot as No. 21, but a different point.
23	1·10	1·3	12·5	0·040	0·060	..	Austenitic manganese-steel casting.
24	0·140	0·010	1·15	0·025	..	..	500 kg. electric steel ingot, rolled.
25	0·290	0·125	0·550	0·016	0·015	..	500 kg. ingot as cast; 0·050 per cent. of aluminium added to melt.
26	0·155	0·010	1·20	0·020	..	..	Same analysis as No. 8; 0·020 per cent. of aluminium added. Rolled.
27	0·065	0·020	0·020	0·090	..	0·150	5 kg. melt made in H.-F. furnace; 0·4 per cent. aluminium added after over-oxidation and before pouring. As cast.
28	0·120	0·140	0·160	0·085	..	0·021	Same remark as for ingot No. 27. Metal not over-oxidised. As cast.
29	0·900	0·210	0·250	..	..	0·005	Same remark as for No. 3.
30	0·185	0·010	1·10	0·025	..	..	500 kg. electric steel ingot, as cast; 0·005 per cent. aluminium added to the melt. Same melt as No. 9.
31	0·185	0·010	1·00	0·016	..	..	Same melt as ingot No. 9; 0·010 per cent. of aluminium added. Rolled.
32	Same	Ingot	as No.	9.			After rolling.
33	0·190	0·010	1·08	0·020	0·018	..	500 kg. ingot; 0·010 per cent. of aluminium added to the melt. Rolled.
34	0·095	0·210	0·510	0·020	0·017	..	Same melt as No. 18 with the addition of 0·010 per cent. of aluminium. As cast.
35	0·080	0·090	0·780	..	..	..	500 kg. ingot, as cast; 0·02 per cent. of aluminium added.
36	0·575	0·095	0·740	..	..	..	500 kg. ingot, as cast; 0·050 per cent. of aluminium added to the melt.
37	0·360	0·300	0·570	0·010	0·015	..	4 kg. ingot taken from an electric furnace before pouring a 15-ton melt.

sulphur and oxygen, are not very numerous, are fairly simple and, hence, well known. The reader is referred on this point to the works of Wohrman (*loc. cit.*), and to the more recent papers of Benedicks and Löfquist (*loc. cit.*), and of Löfquist.[1] It seems difficult to add anything new on this point. However, for the sake of completeness, a few examples of these inclusions will be briefly described. The analyses of the samples examined are set forth in Table I.

A. *Iron Alone.*

(1) *In the Presence of Oxygen.*—The only phase met with in iron in actual practice which contains only very small quantities of concomitant or alloying elements, and especially manganese, is, in the absence of sulphur, the oxide of iron FeO. For this reason, this particular phase is rarely observed, or at least only in isolated cases, in the industrial grades of iron. The oxide FeO is visible in the samples of oxidised metal taken from steel furnaces, when the bath is refined to a sufficient extent to prevent the remaining carbon from reducing it completely during solidification. This oxide may, however, be met with, generally associated with the double oxide of iron and manganese or with sulphides, in the middle region of ingots of rimming steel.

In the metal as cast, the oxide FeO occurs in the form of rounded grains a few microns in diameter, and of rather dark greyish-brown colour (Fig. 1, Plate XXXIX.). Examined by conical illumination (Fig. 2) and at high magnifications, these grains appear perfectly opaque. Immersion in cedar-wood oil considerably reduces their reflecting power. They are generally distributed anyhow, except in certain special cases where they seem to have been forced to the boundaries of the primary crystals and to have a vaguely eutectic character. This fact would appear to show—and it agrees with the results of solubility determinations hitherto made—that FeO segregates to a considerable extent from the liquid mass at the moment of solidification.

(2) *In the Presence of Sulphur.*—It is known that, in the presence of iron alone, sulphur gives rise to a sulphide of iron, FeS. This sulphide is light yellow ochre in colour and has a high reflecting power which is not diminished much by immersion in cedar-wood oil ; it is very opaque, and is only found when the percentage of manganese in the metal is low (less than 0·1 per cent., according to Löfquist). As regards its form, it appears, when present in large quantities, in the form of a network at the joints of the metal grains, or else in irregular grains having more or less rounded contours. The arrangement of the sulphide

[1] Löfquist, *Jernkontorets Annaler*, 1933, vol. 117, pp. 49–111.

inclusions can be modified by heat treatment.[1] When heated for 2 hr. at 1,250° C., a rolled billet containing 0·120 per cent. of sulphur exhibits a complete reorganisation in the orientation of the sulphides (Figs. 3 and 4), which assume a eutectic arrangement, due, probably, to the partial remelting of a very high-sulphur iron/sulphur alloy at the grain-boundaries. FeS also occurs, isolated, in certain unkilled steels, at the centre and the tops of the ingots. As with sulphides generally, it appears to precipitate in the interstices of the dendrites, at a low temperature and towards the end of the period of solidification.

(3) *In the Presence of Sulphur and Oxygen.*—This is the case most frequently met with. In low-alloy iron containing oxygen and sulphur, FeS is often observed in association with FeO. In the liquid state, FeS and FeO are completely miscible and their mixtures solidify with the formation of a two-phase eutectic, the phases being rich in FeO and FeS respectively. For this reason, the inclusions associated with the sulphide and protoxide generally occur in the following forms: (*a*) Globules of FeO surrounded by a crescent of FeS/FeO aggregate of eutectic appearance (Fig. 5); (*b*) globules of FeS surrounded by a eutectic FeS/FeO aggregate (Fig. 6); (*c*) globules of eutectiform FeS/FeO aggregate.

B. *Iron and Manganese.*

(1) *In the Presence of Oxygen.*—The addition of even slight quantities of manganese to oxidised iron has the effect of modifying the colour and transparency of the protoxide inclusions, which retain the same shape, although often slightly more irregular (Fig. 7). The non-metallic phase actually turns darker, passing from greyish-brown to greyish-violet. A blood-red reflection is also noticed in the centre of the coarsest inclusions, especially under oil-immersed magnification. With dark-ground illumination (Fig. 8), the inclusions appear transparent, being red with orange reflections. Since this transparency becomes more pronounced with increase in the percentage of manganese, it seems logical to deduce the presence of a solid solution of MnO in FeO, with increasing MnO contents, in accordance with the FeO/MnO diagram.[2] This phase[3] (Fe, Mn)O is often met with in the form of very fine lines at the grain boundaries (Fig. 9, Plate XL.), or else as a eutectic mass (Fig. 10). With relatively high percentages of manganese (of the order of 1 per cent.), the double oxide becomes still darker and changes shape, becoming irregular and, in certain cases, markedly dendritic (Figs. 11, 12, 13 and 14).

[1] See, recently, Nicdenthal and Bennek, *Archiv für das Eisenhüttenwesen* 1933–34, vol. 7, p. 683.

[2] Andrew, Maddocks and Howat, *Journal of the Iron and Steel Institute*, 1931, No. II. pp. 283 and 309.

[3] See Benedicks and Löfquist, *loc. cit.*

This form of crystallisation indicates that the non-metallic phase has been deposited in the solid state in the liquid metal. The conditions under which the double oxides of iron and manganese form have been specially studied by Körber.[1] The transparency of this phase (dark-ground or immersion objective) remains high and its colour changes to yellowish-green or green.

It is generally admitted that this phase of true green colour is formed by MnO containing very little FeO in solid solution (Löfquist, *loc. cit.*). This hypothesis is very probable, as MnO is green in the free state, and the authors have also met with this same phase in refined ferro-manganese. However, they have not been able to observe the simultaneous presence of red (Fe, Mn)O and green (Mn, Fe)O in one and the same inclusion, such as other writers have reported. The separation of these two phases in the same inclusion also seems to conflict with the appearance of the FeO/MnO diagram (Andrew, Maddocks and Howat, *loc. cit.*), which exhibits a continuous solid solution from FeO to MnO.

MnO also seems capable of assuming a very divided form (Figs. 15 and 16) ; nor is it rare to find a kind of halo of tiny green grains of this same phase surrounding a coarse inclusion of MnO and at a certain distance from it (Fig. 17, Plate XLI.). Sometimes a ring of these tiny grains even surrounds an empty space in which there are no inclusions, slight re-grinding followed by re-polishing causing an inclusion to appear underneath, when the halo assumes a roughly spherical shape. The origin of this curious distribution, which is also found with the silicates and the sulphides, seems difficult to interpret. Examples of this will again be mentioned when dealing with silicate inclusions (Figs. 37 and 39, Plate XLIII.). When the MnO inclusions are comparatively thin and intersected obliquely by the polished surface, the transparency of the inclusions causes interference fringes to appear which, in certain cases, have enabled the refractive index of this phase to be measured.[2] Fig. 18 shows an example of this type of thing.

(2) *In the Presence of Sulphur.*—According to Vogel and Baur,[3] FeS has a very low solvent power for manganese sulphide. MnS, on the contrary, is capable of dissolving sulphide of iron in considerable quantity (55 per cent.). As a matter of fact, when small quantities of manganese are added to the metal bath, or in the presence of small quantities of manganese in steel containing a high percentage of sulphur, a second phase is observed alongside the bright yellow ochre FeS phase. This second phase is scarcely

[1] Körber, *Stahl und Eisen*, 1932, vol. 52, pp. 133–144.

[2] Benedicks and Löfquist, *Jernkontorets Annaler*, 1933, vol. 117, pp. 443–456.

[3] Vogel and Baur, *Archiv für das Eisenhüttenwesen*, 1932–33, vol. 6, pp. 495–500.

any darker, being light greyish-blue in colour. It is probably MnS saturated with FeS (Fig. 19). Whereas FeS is relatively brittle and exhibits polishing pits, this new sulphide is more plastic and can be polished better. If the percentage of manganese is sufficient (about 0·2 per cent.), the FeS phase disappears, to be replaced by this double sulphide. Unlike iron sulphide, it is practically never met with as a network at the grain boundaries, which suggests that this property partly accounts for the effect of the diminution in the red-shortness of iron desulphurised by the addition of manganese. (The study of the malleability of nickel leads to similar conclusions as regards the influence of sulphur, magnesium and manganese).

At relatively low manganese contents, this sulphide is opaque and light coloured. As the percentage of manganese in the steel increases, it becomes darker, assuming the characteristic dove-grey colour. At the same time, it acquires a certain transparency (Benedicks and Löfquist, *loc. cit.*) (Fig 20) and a true yellowish-green to green colour, similar to that of (Mn, Fe)O, but remains more opaque than this oxide.

It should be noted that MnS is not always present in the rounded and more or less irregular form. When present alone, it may also assume the following forms:

(*a*) A eutectic, similar to that already observed for FeS (Fig. 21) (in the vicinity of the discard of large high-manganese ingots). According to Vogel and Baur (*loc. cit.*), this mode of occurrence would be equivalent to a peritectic transformation, a liquid Fe/Mn phase containing sulphur in solution giving rise, on cooling, to another solid Fe/Mn phase containing less sulphur, and to droplets of a (Mn, Fe)S-rich phase which deposit, as formed, at the surface of the metallic crystals in process of growth. This phenomenon would also seem to be connected with a frequent form of coarse globules of sulphide, already referred to (Wohrman, *loc. cit.*) and which contain a secondary metallic inclusion inside them (Fig. 22).

(*b*) A crystallised form. The melting point of pure MnS is round about 1,600° C.—roughly the same as that of iron. If the melting point of the latter is considerably lowered—say, by the addition of carbon—MnS or even (Mn, Fe)S will tend to precipitate in the solid state. This phenomenon is observed in cast irons and also in carbon-steel ingots to which wood charcoal has been added immediately after being cast, as they are thereby highly carburised and also enriched in sulphur by segregation. In the first case, a precipitation of (Mn, Fe)S in crystals (Fig. 23) has been observed, and in the second case, a precipitation in the form of very fine dendrites (Fig. 24, Plate XLII.). Sulphide dendrites may also be observed in cast iron.

(3) *Oxygen and Sulphur.*—It is easy to see that slight additions of manganese to oxidised and slightly sulphurised iron cause the disappearance of a substantial part of the FeS/FeO eutectic border which otherwise nearly always exists around the globules of FeO. This can be interpreted by examining the pure MnO/MnS diagram. Actually, these two substances have a fairly appreciable mutual solubility, and it is probable that the addition of manganese to the FeO/FeS mixtures causes a little sulphide to go into solution again in the oxides, and *vice versa*.

The fine MnO/MnS eutectic can be observed comparatively seldom (Figs. 25, 26 and 27); but the MnO/MnS mixture (or, rather, (Mn, Fe)O/(Mn, Fe)S) is met with in the form of globules of oxide containing a few grains of sulphide (Figs. 28 and 29).

C. *Iron and Silicon.*

In the presence of variable relative proportions of silicon and oxygen, it ought to be possible to retain in steels the inclusions belonging to the SiO_2/FeO series. As a matter of fact, it is difficult to do so in industrial steels and alloys, owing to the almost invariable presence of sufficient quantities of manganese to ensure that the inclusions observed nearly always contain more or less MnO.

It has been possible to study inclusions of silicate of iron alone[1] in light castings specially prepared for this purpose, or in samples taken from steelworks furnaces. The authors prefer to study directly the most general case of inclusions set up by sulphur and oxygen in the presence of manganese and silicon, as such investigation can be carried out on large ingots from current production and, hence, on coarser inclusions. And, as will be seen, from the strictly morphological point of view, MnO and FeO exhibit very similar properties in the presence of SiO_2, and the appearance of iron silicate and manganese silicate inclusions does not differ appreciably from the iron silicate inclusions already described.

D. *Iron, Silicon and Manganese.*

(1) *In the Presence of Oxygen.*—The silicates of iron and manganese are met with in practically all the ordinary carbon steels that have not been deoxidised by aluminium or aluminium alloys. This class of inclusions will be studied by proceeding from the more acid to the more basic types.

A study of the deoxidation of iron by silicon and by manganese has led Körber and Oelsen[2] to believe that, in the absence of

[1] Herty and Fitterer, *Mining and Metallurgical Investigations*, 1928, *Co-operative Bulletin No.* 36; see also, Herty, Christopher and Stewart, *Mining and Metallurgical Investigations*, 1930, *Co-operative Bulletin No.* 38.

[2] Körber and Oelsen, *Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung*, 1933, vol. 15, pp. 271–309.

manganese, slight additions of silicon induced the precipitation of solid silica, which, actually, is undercooled and in the vitreous state. The authors have found the same result in castings of low weight (4 kg.) made in the high-frequency furnace, and, *a fortiori*, for the higher percentages of silicon. By way of example, they remelted and oxidised electrolytic iron, then added 0·5 per cent. of silicon to it. When the metal was cast into a 4-kg. ingot, the total oxygen determination gave :

At the top :	Oxygen,	$0{\cdot}061_8$, $0{\cdot}064_8$ per cent.
In the middle :	,,	$0{\cdot}066_2$, $0{\cdot}067_5$ per cent.
At the bottom :	,,	$0{\cdot}065_2$, $0{\cdot}065_4$, $0{\cdot}062$, $0{\cdot}067_3$ per cent.

which shows that the oxygen is very evenly distributed.

The silica contents, determined by volatilising with chlorine, correspond to oxygen = $0{\cdot}031_8$ and $0{\cdot}033_5$ per cent., and by dissolving in dilute hydrochloric acid to $0{\cdot}036_5$ and $0{\cdot}030_2$, which shows that the inclusions are composed almost entirely of silica.

A microscopic examination reveals spherical globules of variable dimensions fluctuating between one (Fig. 30) and one hundred microns (Fig. 31), whose appearance is well-known and has often been described. They are very transparent, colourless, or slightly yellowish (Fig. 32), and, in polarised light, exhibit the black cross phenomenon (Fig. 33) (Hoyt and Scheil, *loc. cit.*). This latter appearance is not peculiar to vitreous silica, but to all the spherical transparent inclusions, and is probably due to the convergence of the polarised light reflected on the bright ground of the inclusion (Dayton, *loc. cit.*). In the vast majority of cases, these globules of silica are separate, but the authors have been able to confirm their partial coalescence in metal cast very hot and highly oxidised (Fig. 34, Plate XLIII.). (The residue analysis shows that these inclusions are composed of almost pure silica.)

Inclusions of amorphous silica are not very malleable (Fig. 35), and are only slightly deformed, if at all, by rolling.

With high percentages of silicon and average or low percentages of manganese, the products of deoxidation are always rich in silica. However, a certain quantity of manganese and iron goes into solution in the non-metallic substance, in which case acid silicates of iron and manganese occur. For instance, an iron melt containing 0·72 per cent. of added silicon and 0·615 per cent. of added manganese, and averaging 0·054 per cent. of oxygen (vacuum melting method) gave a residue of the following analysis when attacked by chlorine :

	Oxygen due to SiO_2.	SiO_2. %	Oxygen due to FeO.	FeO. %	Oxygen due MnO.	MnO. %	Total Oxygen of the Residue.
No. 1 .	$0{\cdot}048_8$	77·6	$0{\cdot}002_8$	11·0	$0{\cdot}003_0$	11·4	$0{\cdot}054_8$
No. 2 .	$0{\cdot}047_0$	76·7	$0{\cdot}002_8$	11·3	$0{\cdot}003_0$	12·0	$0{\cdot}052_8$

In this case, these inclusions differ very little in appearance from the silica inclusions, being always small, transparent and refracting globules. According to the SiO_2/MnO/FeO diagram published by Körber and Oelsen (*loc. cit.*), when they contain more than 50 per cent. SiO_2 the liquid solutions of silicates of iron and of manganese ought to precipitate solid silica at about 1,600°–1,550° C. on cooling. As a matter of fact, the inclusions in the example selected are of small dimensions, and the rapid cooling prevents their coalescence, for which reason the number of nuclei of crystallisation is small, and the non-metallic substance tends more easily to remain vitreous, *i.e.*, undercooled, the higher the percentage of silica.

Although, in this example, the inclusions all have the same appearance, the composition of the different globules is by no means the same. As a matter of fact, the additions of manganese and silicon were made separately. As regards the silicon, Körber and Oelsen have very clearly suggested that high local concentrations of silicon must cause almost pure silica inclusions to precipitate around the particles of added ferro-silicon, whereas silicates that are clearly more basic can precipitate in the region where the silicon concentration is lower owing to diffusion not taking place instantly. How, then, is one to distinguish between vitreous silica and the basic silicates, as these inclusions, when small, and in the as-cast metal, have a very similar appearance? This can be done by examining them after forging or rolling. The basic silicates (containing less than approx. 50 per cent. of silica) are very malleable, as shown in Fig. 35, which depicts side by side, after the experimental ingot has been forged, an inclusion of silica that is probably almost pure and which is scarcely deformed at all, and a very elongated inclusion of basic silicate.

The analysis given above is certainly only an average general analysis. The same remark applies to large-sized ingots. As a matter of fact, when samples are taken from steelworks furnaces, the presence of small globules of silica can frequently be detected in the metal even a long time after ferro-silicon has been added. The explanation for this is that these inclusions have a high viscosity and a low capacity of coalescence, and they enter into equilibrium with the bath more slowly than the more basic and fluid inclusions do. It is, therefore, fairly safe to say that, every time small, vitreous and only slightly deformed globules are met with in rolled metal, an addition of ferro-silicon has been made to the metal during the process of melting, and probably towards the end.

The inclusions of silicates of iron and manganese met with in ingots of average dimensions (500 kg. and 2 metric tons) alter their appearance when the percentage of silicon in the metal is no longer so high, compared with the percentage of manganese,

for a given percentage of oxygen, and especially when the deoxidation has been effected with silico-manganese. From being perfectly transparent and appearing black in reflected light with a bright point in the centre and a bright annular reflection, they turn a dark greyish-violet and are translucent (Fig. 36). Their true colour is amber yellow. As already indicated, they are very malleable as compared with silica inclusions, and they also possess a strong coalescing power, and hence separate readily. Frequently, accumulations of them are found round about the discard of large ingots, and certain of them may attain dimensions up to a millimetre. The mechanism of this coalescence is still somewhat obscure ; it may be caused either by diffusion between two immediately adjacent inclusions, in which case the non-metallic substance must be to a certain extent soluble in the metal which surrounds it at the temperature involved ; or, as Herty and Fitterer[1] state in a preliminary investigation on deoxidation by ferro-manganese, it may be caused by the relative surface tensions of the inclusions present. (This rôle of the surface tension in the coalescence and elimination of the inclusions during deoxidation by manganese had already been mentioned previously by one of the present authors.[2]) In their paper Herty and Fitterer included a photo-micrograph—unfortunately badly reproduced—representing a large inclusion of silicate of iron and manganese, surrounded by numerous small inclusions. Fig. 37 and Fig. 38 at a higher magnification show a translucent silicate surrounded by inclusions of this kind. Reference to Fig. 39 will show that they may be very small, and appreciably less than 1 micron in size. It has even been possible in certain cases to catch the moment of this coalescence by agglutination (Fig. 40 ; see also Fig. 37), which would militate in favour of this latter hypothesis.

Devitrification, at least to a partial extent, is often observed in the silicates of iron and manganese. This devitrification, which was observed first of all in acid slags and which Whiteley and Hallimond[3] were among the first to mention, takes place in the form of a precipitation of very special structure which the majority of authors agree in identifying with solid silica.[4]

When this precipitation is fine (Andrew, Maddocks and Howat, *loc. cit.*), it has the appearance of a multitude of small globules (Fig. 41, Plate XLIV.) with often slightly angular contours, and darker than the translucent ground of the inclusion. In a more

[1] Herty and Fitterer, *United States Bureau of Mines*, 1931, *Report No.* 3081.

[2] Portevin, *Journal of the Iron and Steel Institute*, 1923, No. II. p. 53.

[3] Whiteley and Hallimond, *Journal of the Iron and Steel Institute*, 1919, No. I. pp. 199–253.

[4] After this research was completed, a paper was published by Urban and Chipman (*loc. cit.*), in which these two authors are said to have been able, by refractive index measurements, to identify this precipitated constituent with cristobalite

advanced state of development, these globules assume the very characteristic "rosette" formation, or a stellate formation proceeding from a centre, like the globulites of the mineralogists (Fig. 42). These "rosettes" may be the microscopic stage in the devitrification of certain glasses. On a macroscopic scale, the rosettes are observed in the form of spherulites, being of similar structure, but having the radiating elements linked up instead of being mostly independent as in the case which is now being dealt with. These rosettes do not seem to be very refractive, and they have a lower reflecting power than the ground-mass, which makes them appear darker than the latter in reflected light, especially at low powers. If the ground-mass is translucent enough (Fig. 42), or, rather, if the inclusion be partially illuminated by diffraction, by means of a narrow eccentric pencil of light (Fig. 43), the shape of these rosettes in space can be very clearly distinguished. Examination with immersion objectives and dark-ground illumination does not give very good results, owing to the refracting powers of the two constituents differing very little from each other (see also Figs. 50 and 51, Plate XLV.). The rosettes may be of different shape. In the majority of small inclusions in which they are present, they take the form of coarsely segmented globules (Fig. 46); whereas, in certain large inclusions, they often tend to assume a markedly crystalline appearance (Fig. 47). Varieties of different appearance may often be met with side by side. Figs. 48 and 49, for instance, show the juxtaposition of a vitreous inclusion, a finely precipitated inclusion, and a "rosette" inclusion.

The ground constituent can be slightly etched by alkaline sodium picrate, but the rosettes cannot (Figs. 44 and 45). This constituent is generally vitreous, but only in a few cases have the authors been able to observe its crystallisation.[1] In certain "rosette" silicates, the ground constituent is more opaque, and light violet in colour. In reflected light, and when examined under a high-power dry objective, it appears heterogeneous (Fig. 50). Examination with an oil-immersion objective reveals a finely eutectiform complex structure (Fig. 51). In the case of pure silicates of iron, without manganese (Herty and Fitterer, *loc. cit.*), and in agreement with the SiO_2/FeO diagram, Herty has already observed similar structures, which are said to be equivalent to a silica/iron-silicate eutectic. The phenomena must be modified by the presence of manganese, but that does not rule out the possibility of this particular structure being a eutectic of silica and a double orthosilicate of iron and manganese, although it is, of course, impossible to be certain on this point.

The mean composition of these silicates has been ascertained

[1] Urban and Chipman, in the paper previously mentioned, were able, though not in every case, to cause devitrification of the ground-mass by suitable annealing.

(by the chlorine method) on a large number of similar samples and has given SiO_2 varying between 45 and 60 per cent. and between 15 and 30 per cent. MnO, the remainder being FeO. These silicates are still resistant enough to the action of chlorine to enable them to be separated by this method, but this composition seems to be a lower limit, because, in the case of certain analyses, the oxygen of the residue tends to be lower than the total oxygen as determined by melting *in vacuo*.

As regards malleability, it has been seen that the moderately basic vitreous silicates were much more malleable than vitreous silica, and the same remark applies to the "rosette" silicates, although the heterogeneity of their structure affects their form after rolling. Fig. 52 (Plate XLVI.) shows the appearance of one of these inclusions after being rolled; its contours are more irregular and the silica rosettes it contains are considerably deformed. This example clearly demonstrates the advantage of studying the initial structure in the metal as cast.

It is somewhat rare to find, in industrial steels, more basic silicate inclusions in which the protoxides FeO and MnO are the main elements. Two examples of such inclusions are given:

(1) In wrought iron, which usually contains very little manganese, silicate inclusions which are not entirely expelled by the forging are frequently met with in the mass. These inclusions, which are very basic, have recently also been mentioned by Löfquist (*loc. cit.*). They are formed by a slightly translucent ground-mass of very fine microcrystalline structure, probably eutectic, and which is difficult to resolve even at high powers (Fig. 53). In this mass are frequently found very well shaped dendrites of greyish-brown iron oxide, which are formed subsequent to the mechanical working of the metal, since they are not deformed. When examined between crossed nicol prisms (Fig. 54), the elements of the ground-mass appear to be crystallised into lamellar groups of yellowish-green colour which do not exhibit double refraction. The ferrous oxide appears completely opaque and optically inactive. According to the FeO/SiO_2 diagram, this ground-mass ought to be constituted by the eutectic $FeO/2FeO.SiO_2$ (ferrous oxide/fayalite).

(2) In a very mild unkilled steel containing very little silicon as compared with relatively high manganese and oxygen, a large number of globular inclusions of the type shown in Fig. 55 have been found. The ground-mass is also finely eutectic; the primary constituent segregated is (Fe, Mn)O of true reddish-brown colour. A third constituent—a sulphide—is also visible, and reference will be made to this later on.

(2) *In the Presence of Sulphur, and* (3) *In the Presence of Sulphur and Oxygen*.—In the absence of oxides, the relative percentages of sulphur and manganese alone govern the appearance

of the sulphides met with, as the silicon content does not appear to play any part.

The phenomena are different when oxygen is present, in sufficient quantities, as a supplementary element ; sulphides and silicates may actually be present in the same metal, either separately or in association.

The researches carried out on the equilibrium diagrams of silicates/sulphides of iron and manganese, especially the recent work done in Great Britain by J. H. Andrew and his collaborators,[1] and in Soviet Russia by Ginzberg and his collaborators,[2] have established the following facts :

(*a*) In the liquid state, silicates and sulphides give liquid solutions, either homogeneous throughout the series[1] ($MnS/MnO.SiO_2$, $MnS/2FeO.SiO_2$), or exhibiting a miscibility gap in a certain region[2] ($FeS/MnO.SiO_2$ and $FeS/2FeO.SiO_2$).

(*b*) At solidification and at the equilibrium point, there is a sulphide-silicate eutectic for each of the systems mentioned.

(*c*) In the solid state, each of these phases is capable of containing a certain quantity of the other in solid solution.

It is possible to recognise, at least partially, structures whose appearance is connected with the properties just mentioned, when examining the mixed silicate-sulphide inclusions in steels containing silicon and manganese.

(*a*) The best known appearance of the mixed silicate/sulphide inclusions is that of the inclusions which are termed " duplex," and of which Fig. 56 shows an example very clearly. Droplets of sulphide (Mn, Fe)S are irregularly arranged against the inside wall of a globular inclusion of vitreous silicate which, as may be seen by subsequent rolling (see Fig. 103, Plate LV.), is very malleable, and, hence, relatively basic. Similar droplets are also visible in the inclusion of Fig. 55. It is not difficult to account for a structure of this kind if it is assumed that, at the moment of its formation, the silicate inclusion was able to contain, in the liquid state and in solution, a certain quantity of sulphide which segregated while still in the liquid state, owing to the existence of a miscibility gap, during the period of cooling following its formation (Benedicks and Löfquist, *loc. cit.*)

(*b*) The silicate/sulphide eutectics are rarely met with as constituent elements of the inclusions. Generally speaking, the mixed inclusions exhibit only two phases, which are clearly separated into relatively coarse elements. However, the authors have been able to identify these eutectics in a few cases :

(1) An extra-mild steel, highly oxidised and manganiferous exhibited numerous inclusions having, at low powers, the structure

[1] Andrew, Maddocks and Fowler, *Journal of the Iron and Steel Institute*, 1931, No. II. pp. 295–325.

[2] Ginzberg, Selivanow and Nikolsky, *Journal of the Institute of Metals, U.S.S.R.*, 1931, pp. 25–29 and 74–78.

shown in Fig. 57 (Plate XLVII.), *i.e.*, a constituent of average greyish-violet colour, roughly globular in shape, surrounded by a ground-mass of the same tint, but darker. By dark-ground illumination (Fig. 58), the ground-mass constituent was light green and very transparent, whereas the separate constituent was more opaque and its true colour bottle green. The relative colours of the oxide and the sulphide of manganese, as well as their relative transparency would, on a superficial examination, have led the observer to identify them as globules of sulphide in a matrix of oxide. Examination with an oil-immersion objective, however, showed that the ground-mass had a fine-grained eutectic structure (Fig. 60), while even weak etching with a 10 per cent. aqueous solution of chromic acid (to bring out the sulphides of iron and manganese (Campbell and Comstock, *loc. cit.*)) completely blackened the ground-mass (Fig. 57), whereas the light constituent was only dissolved by subsequently attacking it with a saturated solution of alcohol and stannous chloride (to bring out the oxides of iron and manganese). An examination of other similar inclusions by a narrow beam of light enabled the structure of the ground-mass to be defined, this being formed by small green globules which are more or less opaque and surrounded by a highly transparent vitreous mass (Fig. 61). Probably it is these tiny grains, consisting, very likely, of sulphide of manganese, which caused the apparent blackening of the ground-mass shown in Fig. 59, and probably the mass surrounding them is a vitreous basic silicate of manganese whose primary constituent is the protoxide, MnO.

(2) "Duplex" inclusions of the type shown in Fig. 62 have been observed in a large ingot of unkilled mild steel, in a region particularly rich in sulphides. The phases present are :

Manganese sulphide separated into very clearly defined globules. Curiously enough, there appear to be two sulphide phases, (*a*) a light grey one present in large quantity, and (*b*) a dark grey one present in a lesser quantity ; both phases can be attacked by chromic acid, and are practically opaque.

A constituent of lamellar structure and eutectic ground-mass, which may probably be a sulphide/silicate of iron and manganese eutectic. Examination in polarised light (Fig. 63) shows its relative transparency as compared with the primary segregated sulphides.

(3) Mixed silicate/sulphide inclusions may frequently be found in 13 per cent. manganese steel for castings, which, compared with mild steels, contain a large percentage of sulphur. Fig. 64 shows the appearance of a coarse inclusion met with in a casting. The primary dendritic constituent is pure sulphide of manganese, as it is probable that, in the presence of these percentages of manganese, the inclusion must contain only very little iron. The basic constituent, greenish-brown and more transparent

throughout (Fig. 65) has a lamellar orientation at low powers, and exhibits, under an immersion objective, a multitude of small globules of MnS in a mass of silicate. It is interesting to compare this appearance with the aspect of Figs. 43 and 44 from the research of Andrew, Maddocks and Fowler (*loc. cit.*), which, according to these writers, represents MnS in a matrix of eutectic $MnS/MnO.SiO_2$. There is a great similarity in the structure.

(*c*) Several writers agree that the silicate inclusions contain sulphide in solution and *vice versa*. From the morphological point of view, this hypothesis is usually not very important, since the associated sulphides and silicates of iron and manganese do not differ in appearance from (1) the silicates which may be met with in synthetic castings containing very little sulphur, and (2) the sulphides met with in castings of the same type and of very low silicon content. In this connection and in support of this hypothesis, the authors may merely mention a fact of everyday observation, which, however, has only a qualitative value. It is this : For the same percentage of sulphur as ascertained by analysis, a steel seems to exhibit, micrographically, more visible sulphides the more carbon it contains ; the same applies when aluminium has been added to it. Actually, in the former case, the total oxygen is usually low ; while in the second case, as will be seen later on, the oxide inclusions consist of alumina. In both cases, the formation of silicates is prevented, and the sulphides remain out of solution. This is only supposition, of course, since it is equally possible that the aluminium and the carbon diminish the solubility of the sulphur in solid iron, and, as has been seen, this solubility does not seem to be negligible.

The " duplex " inclusions afford a means of assessing the relative malleability of their constituents. As may be seen from Fig. 67 (Plate XLVIII.), the basic vitreous silicates seem to be generally more malleable than the sulphides and are more drawn out than the latter by a moderate amount of hot-working.

E. *Iron (Manganese), and Aluminium.*

(1) *In the Presence of Oxygen.*—It is known that aluminium has a very great affinity for oxygen,[1] with which it gives alumina. In ordinary carbon steels, even those containing manganese, if the quantity of aluminium added as a deoxidising agent is high compared with the state of oxidation of the metal, alumina forms almost exclusively.

The alumina inclusions are most frequently present in the form of grains of small dimension (1 to 5 microns), either isolated, or else in the very characteristic form of clusters of very variable total dimensions (Figs. 68 and 69), which are elongated by rolling

[1] *See*, for instance, the recent paper by Chipman, *Transactions of the American Society for Metals*, 1934, vol. 22, pp. 422–425.

(Figs. 70 and 71). These grains are transparent, irregular or roughly crystalline, colourless to violet-black in reflected light, and of true yellowish-white colour when examined by dark ground illumination. They are very refractive, difficult to polish because they are very hard (corindon), and are very often accompanied, in the direction of polishing, by a "comet's tail" streak. A coarser and more irregular grain is occasionally met with among the small particles of alumina (Fig. 70). In all cases where these inclusions alone are observed, the total oxygen ascertained by melting *in vacuo* is found to be present as alumina in the residue left when they are attacked by dilute hydrochloric acid.

The alumina inclusions do not always exhibit the same external appearance, as this probably depends on the conditions under which they are formed.

Suppose solid metallic aluminium is added to a bath of steel which is normal as regards temperature and degree of oxidation. In the regions where the aluminium has a sufficiently high degree of concentration, the same phenomenon will occur as the one mentioned when dealing with silicon, except that it will be more pronounced owing to the affinity of aluminium for oxygen; and the oxygen contained in the metal, say, in the form of FeO, will, in this region, be instantaneously converted into solid alumina, whose melting point is 2,050° C., or much higher than that of steel. This solid alumina has crystalline contours which are usually badly defined—and this is the form most generally met with.

Suppose the steel is highly oxidised and superheated, and that there is only a small quantity of it; let a considerable percentage of aluminium be added to it. The local rise of temperature, due to the heat of reaction, may be sufficient for the alumina to form in the liquid state. The following experiment proves this point: Electrolytic iron was melted in a high-frequency furnace, superheated and super-oxidised (oxygen, 0·150 per cent. approx.), then 0·400 per cent. of aluminium was added to it. The alumina inclusions observed were very dense and numerous, and all of them had the globular vitreous appearance (Fig. 72, Plate XLIX.), which tends to prove that they formed in the liquid state. They seemed more refractive than the silica inclusions of the same size (n_{SiO_2} = approx. 1·5; $n_{Al_2O_3}$ = 1·77). Needless to say this form is rarely met with under practical conditions.

Finally, alumina may be met with in the form of small and well-defined crystalline particles, although it is impossible to define the conditions which lead to these regular external shapes being obtained.

(*a*) A melt of electrolytic iron containing a low percentage of oxygen (oxygen, 0·021 per cent.) was deoxidised, under the same conditions as above, by 0·400 per cent. of aluminium. The

inclusions observed frequently exhibited the crystalline appearance shown in Fig. 73. Observed in polarised light under crossed nicol prisms, the majority of these small crystals exhibit two badly defined positions of extinction, owing to their uneven thickness and the reflections from the ground-mass of the inclusions.

(*b*) In a 4-kg. ingot taken from a steel furnace after aluminium was added (unalloyed carbon steel), the alumina inclusions were present in the form of particles crystallised into small rods and small crystals (Figs. 74 and 75) grouped together. The total oxygen as ascertained by vacuum melting was 0·005 per cent., and the oxygen of the residue obtained after treatment with hydrochloric acid, and composed solely of alumina, was $0 \cdot 005_7$.

What takes place when, to a metal containing only a small percentage of silicon, a quantity of aluminium is added which is small compared with the oxygen which it contains ?

Certain writers, particularly Herty and his collaborators,[1] in their research on deoxidation by aluminium, and Löfquist (*loc. cit.*) have, in this connection, mentioned the existence of inclusions which are often crystalline, of larger dimensions than the average alumina inclusions, often darker-coloured and less transparent than the latter, and which are comparable to the protoxide inclusions (Fe, Mn)O. These particular inclusions they regard as aluminates of iron.

This hypothesis is admissible. As a matter of fact, Herty gives analyses of residues obtained by Dickenson's method which, as is known, does not enable the free protoxides to be recovered, and these residues contain large proportions of FeO besides alumina. But it appears to the authors open to discussion to affirm, from the appearance of the photomicrograph, that a definite combination is present between the alumina and the protoxides and to go so far as to define it, for instance, as spinel (Mn, Fe)O.Al_2O_3. The following are the reasons for this objection :

(*a*) As stated above, the affinity of aluminium for oxygen and the stability of alumina when formed are such that it seems inevitable that even a small addition of aluminium must cause many types of inclusions to form in steel which is incompletely killed ; there will be alumina at places where the local concentration of aluminium is average or relatively high ; perhaps aluminates at points where the aluminium concentration is low, plus the protoxide in excess. It is therefore impossible to determine the composition of isolated inclusions met with by analysing the residues.

(*b*) On the other hand, the mere external appearance of the crystalline inclusions observed cannot provide accurate data

[1] Herty, Fitterer and Byrns, *Mining and Metallurgical Investigations*, 1930, *Co-operative Bulletin No.* 46.

as to their crystalline system, because (i) the observer does not know the orientation of the plane of the micro-section with reference to the faces of the crystal cut through; and (ii) the original theoretical shape of the crystal may be affected by a large number of bounding surfaces and correspond to multiple habits. This does not, of course, include any allowance for the lack of accuracy caused by the rounding of the angles, &c. Whereas the mineralogist works on thin plates of minerals of known analysis and having constituents of large dimensions, it is impossible in the present case to fix the position of the external faces with reference to the axes of the crystals studied, with the present technical methods available for studying inclusions. What observation in polarised light does enable one to do, however, is to distinguish between isotropic crystals, which do not exhibit any variations in brightness, and which therefore belong to the regular system, and the anisotropic crystals, which exhibit positions of extinction, usually poorly defined (due to stray reflections on the background of the crystals), and belong to one of the other systems.

That is why, even presuming the existence of aluminates of iron and manganese to be probable in the form of inclusions, there is uncertainty as to their precise character.

(2) *In the Presence of Sulphur, and* (3) *In the Presence of Sulphur and Oxygen.*—Although aluminium sulphide, Al_2S_3, has a very high heat of formation, it has not been possible to isolate it previously in the form of non-metallic inclusions. To confirm this fact, the authors prepared, in the high-frequency furnace, a melt of 4 kg. of electrolytic iron to which 0·1 per cent. of sulphur and 0·4 per cent. of aluminium were added.

The sulphide inclusions observed presented the appearance of sulphide of iron from the morphological point of view as well as in their etching characteristics. In cases where appreciable quantities of alumina are visible, and for obscure reasons, which may be of a physico-chemical or mechanical nature (Löfquist, *loc. cit.*), these inclusions, when observed, are frequently surrounded by a mass of sulphide, giving an appearance which is sometimes of a eutectic nature (Fig. 76).

F. *Iron, Manganese, Silicon and Aluminium.*

In the Presence of Sulphur and Oxygen.—The most general case—that of unalloyed carbon steel containing silicon and manganese and with aluminium additions—will be first dealt with. From the morphological standpoint, two cases may be differentiated:

(*a*) Alumina inclusions are not observed in the metal, but definite silico-aluminates are.

This case arose in a series of 500-kg. ingots of unkilled and semi-killed steel of low silicon (0·010 per cent.) and high manganese (1 per cent.) contents, and to which very small quantities of aluminium of the order of 0·005 to 0·010 per cent. were added. By examining the tops of these ingots (and, more particularly, the crop-end or discard) as cast, then as rolled, it was possible to find a series of large separated inclusions in which the transformations caused by adding the aluminium were progressive.

In the initial stage, the inclusions, which are globular, pinkish-violet in reflected light, amber-yellow under conical illumination (Fig. 77), closely resemble the coarse, vitreous basic silicates of the very malleable type described in section *D*. In the next stage, a very fine precipitation of small crystalline constituents is noted in the midst of the vitreous mass. These constituents are very little darker than the ground-mass but are harder, so that by prolonged polishing they can be given a certain relief with respect to the rest of the inclusion. They are best examined by prism illumination (slightly oblique illumination) and by closing the illuminator diaphragm to its smallest extent so as to give the maximum relief effect. The photomicrograph (Fig. 78) brings out very clearly the square shape of the coarser of these small crystals and the shadow it casts. In another inclusion (Fig. 79), where precipitation is already very pronounced, the crystals show up very clearly and are octahedral in appearance. Their true colour is brick-red to purple-red.

Fig. 80 (Plate L.) shows the appearance of a very large crystal of these inclusions found in the actual crop-end of the ingot. The abundant precipitation of this constituent has taken place in the form of groups of crystals, which are often twinned (note the large crystal joined to a smaller one at the top left-hand portion of the inclusion). The ground-mass, dark pink in colour in reflected light (Fig. 81), and whose true colour is pale yellow, has also crystallised in the form of long needles or lamellæ of dendritic structure (Fig. 82). It is easily attacked by alkaline sodium picrate.

A very interesting structure is that exhibited by the inclusion shown in Fig. 83 (Plate LI.). In this inclusion, the matrix of which has remained vitreous, will be seen a very large single crystal twinned with a smaller one. This crystal, which is very clearly defined, is bounded towards the right by the actual wall of the inclusion, which is irregularly spherical. This would seem to prove that the crystal precipitated in the actual mass of the liquid inclusion when the latter formed, and that it grew from a single nucleus situated on the wall of the non-metallic drop, or was simply arrested by this wall. The decrease in opacity towards the lower left-hand corner, which is visible under dark-

ground illumination (Fig. 84), shows that the polished surface passes approximately through one of the corners of the crystal. No positions of extinction are observed in polarised light, which indicates that this constituent is isotropic.

On a sample of rolled metal manufactured under the same conditions, but containing a slightly higher percentage (0·010 per cent.) of added aluminium, these crystals formed, excluding the vitreous matrix, a large part of the mass of the inclusions, and were also accompanied by clusters of alumina particles (see farther on). Analyses of the residues by the chlorine method gave the following results :

Analysis No. 1 : SiO_2, 26·4 ; Al_2O_3, 43·6 ; MnO, 16·0 ; FeO, 14·0 per cent.
Total oxygen of residue : $0·012_2$ per cent.

Analysis No. 2 : SiO_2, 25·2 ; Al_2O_3, 46·8 ; MnO, 16·8 ; FeO, 11·8 per cent.
Total oxygen of residue : $0·013_1$ per cent.
Total oxygen by melting *in vacuo* : $0·012_7$, 0·014 per cent.

It is highly probable, therefore, that the silica and the protoxides come exclusively from the crystalline constituent observed, the alumina being in excess in the form of clusters of free alumina, and no other oxide being visible. This constituent ought, therefore, to be a silico-aluminate of iron and manganese in accordance with the formula :

$$p\mathrm{SiO_2}.m\mathrm{Al_2O_3}.p(\mathrm{Fe,Mn})\mathrm{O}.$$

Owing to the presence of free alumina, it is unfortunately impossible to determine the ratio m/p by chemical analysis.

A mineralogical substance which might agree with this composition would be a garnet of iron and manganese belonging to the almandine-spessartine series, of composition $3SiO_2.Al_2O_3.3(Fe, Mn)O$. This mineral belongs to the regular system, its true colour is reddish-brown to garnet red, and it usually crystallises in dodecahedra and octahedra.

The authors thought it would be interesting to make rolling tests on an ingot containing an exceptional quantity of the inclusions described above, to see how they deform and to ascertain the malleability of their different constituents.

Fig. 86 (Plate LII.) shows a large inclusion of the type depicted in Fig. 80, but after rolling, the degree of hot-working being approximately 10. It will be noted that, at low powers, the structure seems to remain practically the same, large crystals of silico-aluminate, more or less deformed and broken, being encased in a vitreous matrix. At higher powers and with oblique illumination (Fig. 85, Plate LI.), their characteristic arrangement in groups of small crystals, accompanied by numerous coarser constituents, will be noted. At still higher powers (Fig. 87),

small light grey dendrites, which seem to be formed of sulphide of iron and manganese, will be seen in the ground-mass between the crystals. Examination with an oil-immersion objective and under a narrow pencil of light shows that these dendrites consist of a transparent and refringent mass of true light-green colour. In this case, it is possible to focus either on the front branches (Fig. 91, Plate LIV.) or the side branches (Fig. 92) of the dendritic structure.

Light etching with dilute chromic acid completely dissolves these dendrites (Fig. 88), which certainly tends to show that they were formed by a high-sulphur phase. Their formation very probably took place during the cooling which followed rolling, for they do not seem to have been affected by the process of deformation, otherwise they would have segregated in the solid state from a pasty or partially molten mass. Further, etching with sodium picrate considerably attacks the ground-mass (Fig. 89) (a characteristic of the basic silicates), which is completely and rapidly dissolved by a short period of etching with hydrofluoric acid (Fig. 90). This latter etching also extends to the surface of the silico-aluminate crystals, which it roughens by corrosion. It should then be noted that the crystals exhibit numerous bright reflections ; for, as the intercrystalline substance has been dissolved, total reflections are produced on the far faces of the crystals.

The complete examination of this large inclusion therefore shows (*a*) that the malleable ground-mass " protects " the crystals of silico-aluminate (which cannot be deformed easily) from the crushing caused by rolling ; (*b*) that the basic silicate dissolves—at all events at temperatures of the order prevailing in rolling operations—a certain quantity of sulphur, which precipitates partially when cooling.

The appearance of the inclusions will therefore vary after rolling, depending on the quantity of malleable vitreous substance surrounding the silico-aluminate crystals. Figs. 93 and 94 show the appearance of a rolled inclusion containing very few crystalline constituents. Its general appearance is smooth-edged. Fig. 95, on the other hand, shows the effect of rolling on an inclusion in which the influence of the aluminium was sufficient to transform the whole mass into silico-aluminate, the inclusion being less elongated and having uneven edges. In this case, the quantity of binder was just enough to allow the crystals, during the process of deformation, to slide on each other without breaking, and to retain their original very clearly defined shapes (Fig. 96). An interesting effect was produced by lightly etching this inclusion with hydrofluoric acid ; the whole of the vitreous mass forming the optical and mechanical bond having been dissolved, the inclusion appeared to be filled with grains of refringent sand

(Fig. 97). Actually, these grains have no cohesion, and the mere action of lightly passing a brush or a cloth over the specimen is sufficient to wipe out what is on it.

When the ground-mass is absent, the crystals assume the appearance of a string of beads, and are deformed or broken, but are easily distinguishable from the alumina by being softer, true yellow to red in colour, and having crystalline facets (Fig. 98, Plate LV.).

Finally, where the action of the aluminium was more energetic still, owing to a higher local concentration, the silico-aluminate type changes to the alumina type, the phases being sometimes exhibited in the microscope field of view together (Fig. 99).

The ordinary "duplex" inclusions are modified only in the silicate phase. Fig. 100 shows an inclusion which, at first sight does not appear to differ from them. However, by focusing on the ground-mass of the inclusion, a very clearly defined crystal of silica-alumina may be observed alongside a spherulite of sulphide (Fig. 101). If the conversion into silico-aluminate is complete, the sulphide is expelled from the transformed inclusion, but may remain mechanically associated with the crystals of silico-aluminate, probably owing to its surface tension (Fig. 102).

Fig. 103 gives a very clear idea of the relative malleability of the three phases after rolling, the following appearing in increasing order: The undeformed silico-aluminate, the moderately deformed sulphide, and the very malleable vitreous silicate.

(*b*) The second case to be considered is that in which the intermediate phase does not occur, and in which alumina, or a compound very rich in alumina, forms immediately.

This is also the case with carbon steels melted or deoxidised with manganese and silicon so that the products of deoxidation are silicates which are relatively acid as compared with those of the previous example ("rosette" silicates, for instance), and to which increasing additions of aluminium have been made.

As a matter of fact, it is possible, as will be seen below, to establish, as in case (*a*), different morphological stages of the inclusions met with.

Fig. 104 (Plate LVI.) shows a translucent globular silicate containing a few grains of alumina, which are recognisable by their hardness, as they are hard to polish (*cf.* the peripheral groove surrounding each alumina grain), their transparency in a conical beam of light (Fig. 105) and their true pale-yellow colour.

Identically the same appearance is met with in very large inclusions, probably exogenous or mixed in character, found in basic or acid steels. On a very large inclusion of this type found in the metal as cast, and to which aluminium had been added, the authors were able to photograph this very characteristic "rice grain" appearance of the particles of alumina surrounded

by a vitreous mass (Fig. 106). Where a fairly large amount of aluminium has been added, it is possible (as in Fig. 107) to see how the grains of alumina practically fill up the entire space. Some of these grains have also a pronounced crystalline appearance. One fact nearly always observed is that there are metallic globules, which are more or less coarse and irregular, in the middle of the inclusion.

Very dense and numerous clusters of alumina are also frequently found in the vicinity of these inclusions (Fig. 108).

In concluding, the authors would like to mention the case where the inclusions met with do not seem to contain either iron or manganese, but would appear to consist of a silicate of alumina. Löfquist (*loc. cit.*) has found, in a high-silicon steel, small crystalline inclusions of rhombic shape which he seems to think are mullite. The residue from the electrolytic attack of the sample contains them, corresponding to the formula $3Al_2O_3 . 2SiO_2$ with a slight excess of silica. That author does not mention, however, whether the sample contained only inclusions of this type. In a 4-kg. ingot taken from a steelworks furnace during a medium-carbon steel heat, and after aluminium had been added, the present authors have been able to observe well defined, birefringent and perfectly colourless crystalline constituents accompanying inclusions of alumina (Figs. 109 and 110). An analysis of the residue from the reaction gave a composition equivalent to $4 \cdot 7\ Al_2O_3 . 2SiO_2$, excluding FeO and MnO. The total oxygen of the residue was $0 \cdot 009_6$ per cent., and the total oxygen by the vacuum melting method $0 \cdot 009_4$ per cent. It is quite probable therefore, that these are silicates of alumina, although it is impossible to identify the compound with certainty.

Conclusions.

This first paper is a brief review of the morphological aspects of the non-metallic inclusions, of which it demonstrates the multiplicity, even when observations are confined to the relatively simple case of unalloyed carbon steels. The authors do not claim, however, that this morphological study is complete ; it is certain that, even in the compass of the first part of this paper, other types of inclusions may be met with.

The paper does enable certain general observations to be made regarding the crystallography or appearance of the inclusions and their relation to the composition and method of manufacture of the metal, besides which it throws a certain amount of light on certain features of the manufacturing process, especially as regards the conditions under which the inclusions are formed.

(1) The appearance of the inclusions is an extremely useful factor to consider in connection with their determination, but

is by no means entirely specific. Although, as in the identification of minerals and rocks, it may be a quick and to a certain extent instantaneous means of enabling a skilled and trained observer to identify the nature of the constituents in certain specific cases, it should be noted that this method of identification depends not only on the *nature* of the inclusions, but on many other factors, such as: The temperature of formation, the properties of the inclusion (especially its viscosity and surface tension), degree of mixing, state of the metal (liquid in process, or at the end of solidification), origin (precipitation within the metal, or from an existing inclusion by reduction of the latter), &c.

To mention merely one simple and known example relating to the metal as cast, the high manganous sulphide inclusions, commonly termed sulphide of manganese, may be of crystalline appearance (polyhedral or dendritic) or globular, depending upon whether they form in the liquid metal (carburised cast irons or steels) or during the process of solidification (mild steels).

Generally speaking, one might be tempted to say that an idiomorphous appearance corresponds to a crystalline formation in a liquid metal, and a globular appearance to a liquid or vitreous state of the inclusion at the moment of formation, but this must not be taken as an absolute rule.

In the micrographic identification of the inclusions, the physical processes (examining the colour, the sheen and the transparency by different modes of illumination) play just as important a part as the chemical processes (etching by reagents).

(2) As already stated, during the process of steel manufacture an "evolution of the inclusions" is found conjointly with the chemical modifications in the metal due to the tendency of the *metal/inclusion* equilibrium to be set up, the latter playing the part of an internal micro-slag. In many cases, however, the reactions are incomplete and even irreversible, as, for instance, in the formation of alumina from the metallic phase or from inclusions already existing.

By attentively studying the inclusions in the metal, the skilled and conscientious observer is able to reconstruct certain "stages" or retrace certain operations performed during the manufacture of the metal, and which of course elude all other methods of investigation. These operations leave indelible traces on the nature or appearance of the inclusion. The greatest prudence must, however, be exercised when pronouncing judgment on these inclusions, as the large number of different inclusions met with in steel makes it difficult to arrive at a true assessment. This also demonstrates the irregular nature of the action of the additions, which cause localised increases of concentration and also of temperature, the latter being revealed by the melting of certain refractory inclusions.

The inclusions are much more difficult to identify in forged metal, in which they lose their typical appearance. On the other hand, this examination gives information as to the plasticity and hot-shortness of the inclusions.

This research will be completed by a second portion dealing with the morphology of the inclusions in the iron alloys containing special elements, and chromium in particular.

This investigation has been carried out in the oxygen and inclusion estimating department of the Ugine Electrical Steel Works. The authors thank the management of these works for permission to publish these results, and particularly Monsieur Perrin for his kind assistance in the work.

Description of Illustrations on Plates XXXIX.—LVI.

Plate XXXIX.

Fig. 1.—(Sample No. 1). Glass plate illumination. Oil immersion. Globules of ferrous oxide, FeO. × 1,650.

Fig. 2.—(Sample No. 1). Conical illumination. Oil immersion. Same point as Fig. 1. The inclusions are markedly opaque. × 1,650.

Fig. 3.—(Sample No. 2). Prism illumination. Network of sulphide of iron in a high-sulphur and low-manganese steel. × 165.

Fig. 4.—(Sample No. 2). Glass plate illumination. Same metal as No. 3, after heat-treatment. The partial remelting of the sulphide eutectic has displaced the network, the sulphides having assumed a eutectic arrangement. × 65.

Fig. 5.—(Sample No. 3). Glass plate illumination. Oil immersion. Globules of ferrous oxide surrounded by an interrupted crescent of characteristic appearance, of protoxide-sulphide of iron. × 1,100.

Fig. 6.—(Sample No. 4). Prism illumination. Oil immersion. Globule of sulphide of iron accompanied by protoxide-sulphide eutectic. × 2,200.

Fig. 7.—(Sample No. 5). Glass plate illumination. Alignment of grains of protoxide (Fe, Mn)O in a slightly hot-worked steel. Some of these grains exhibit orange-coloured reflections at the centre. × 660.

Fig. 8.—(Sample No. 5). Conical illumination. Same point as Fig. 7. All the inclusions are translucent and, according to their thickness below the plane of polishing, have a colour varying from orange-yellow to blood-red. × 660.

Plate XL.

Fig. 9.—(Sample No. 6). Glass plate illumination. Lines and eutectic areas of protoxide of iron and manganese. × 330.

Fig. 10.—(Sample No. 6). Glass plate illumination. Rounded eutecticform mass of globules of protoxide of iron and manganese. × 550.

Fig. 11.—(Sample No. 7). Glass plate illumination. Irregular inclusions of protoxide of iron and manganese. × 330.

Fig. 12.—(Sample No. 7). Glass plate illumination. Oil immersion. Three crystals of protoxide of manganese and iron, rich in manganese, which are probably the section of a similar dendritic crystal. The central reflection of each crystal is greenish-yellow. × 1,100.

Fig. 13.—(Sample No. 7). Glass plate illumination. Oil immersion. Crystal of (Mn, Fe)O in the form of a cross. The arms of the cross are very slender, being translucent and pale green. × 2,200.

FIG. 14.—(Sample No. 7). Glass plate illumination. Oil immersion. Incipient dendrite of (Mn, Fe)O. × 1,100.

FIG. 15.—(Sample No. 7). Glass plate illumination. Inclusions of manganous oxide in a very divided form. × 330.

FIG. 16.—(Sample No. 7). Conical illumination. Same point as Fig. 15; the powdery inclusions may clearly be seen surrounding the spaces which are devoid of them. × 330.

PLATE XLI.

FIG. 17.—(Sample No. 7). Glass plate illumination. Group of irregular inclusions of protoxide of manganese (and of iron), more or less joined together, with some metal entrained. This area is surrounded by a kind of ring of small inclusions of the same nature. × 165.

FIG. 18.—(Sample No. 7). Glass plate illumination. Oil immersion. Irregular translucent inclusion of MnO, having interference fringes at its edges. × 1,100.

FIG. 19.—(Sample No. 6). Glass plate illumination. Oil immersion. Bright yellow inclusion of sulphide of iron containing an elongated globule of manganese/iron sulphide of bluish-grey colour. × 2,200.

FIG. 20.—(Sample No. 8). Glass plate illumination. Oil immersion. Translucent elongated inclusion of sulphide of manganese and iron. (The axial reflection is yellowish-green.) × 1,100.

FIG. 21.—(Sample No. 9). Glass plate illumination. Eutectiform group of small inclusions of sulphide of manganese and iron, found at the axis of the crop-end of an ingot. × 330.

FIG. 22.—(Sample No. 9). Glass plate illumination. Globules of sulphide of manganese and iron containing a secondary metallic inclusion. × 660.

FIG. 23.—(Sample No. 10). Glass plate illumination. Inclusion of crystallised sulphide of manganese and iron, in a cast iron. × 660.

PLATE XLII.

FIG. 24.—(Sample No. 11). Glass plate illumination. Dendritic inclusion of manganese sulphide in the crop-end (discard) of an ingot. × 660.

FIG. 25.—(Sample No. 6). Glass plate illumination. Oil immersion. Eutectic inclusion (Mn, Fe)S/(Mn, Fe)O. Sulphide light, protoxide dark. × 2,200.

FIG. 26.—(Sample No. 6). Glass plate illumination. Oil immersion. Same inclusion as Fig. 25. Slightly etched with chromic acid. Only the sulphide is attacked. The appearance seems to be the negative of that shown in the previous illustration. × 2,200.

FIG. 27.—(Sample No. 9). Glass plate illumination. Oil immersion. Elongated inclusion comprising four globules of irregular shape, of sulphide (Mn, Fe)S, connected by the eutectic sulphide-protoxide of manganese and iron. × 1,100.

FIG. 28.—(Sample No. 6). Glass plate illumination. Oil immersion. Mixed oxide/sulphide inclusions, one having a roughly eutectic structure. In the large inclusion, the protoxide is translucent. × 2,200.

FIG. 29.—(Sample No. 6). Glass plate illumination. Oil immersion. Same inclusions as Fig. 28, after lightly etching with chromic acid. Only the sulphide is attacked. × 2,200.

FIG. 30.—(Sample No. 12). Glass plate illumination. Oil immersion. Tiny inclusions of vitreous silica. × 2,200.

FIG. 31.—(Sample No. 12). Glass plate illumination. Globule of vitreous silica. × 330.

FIG. 32.—(Sample No. 12). Conical illumination. Same point as Fig. 31, showing the transparency of the inclusion. × 330.

FIG. 33.—(Sample No. 12). Polarised light, crossed nicols. Same point as Fig. 31. Black cross. × 330.

Plate XLIII.

Fig. 34.—(Sample No. 13). Prism illumination. Inclusions of vitreous silica in process of coalescence. × 660.

Fig. 35.—(Sample No. 14). Glass plate illumination. Two inclusions, one of vitreous silica, the other of more basic silicate, after hot-working. Only the silicate inclusion is elongated in the direction of rolling, the silica inclusion having undergone only an inappreciable elongation. × 165.

Fig. 36.—(Sample No. 15). Glass plate illumination. Inclusion of slightly translucent silicate. × 330.

Fig. 37.—(Sample No. 15). Glass plate illumination. Translucent silicate inclusion, surrounded by a ring of small vitreous inclusions. × 330.

Fig. 38.—(Sample No. 15). Glass plate illumination. Oil immersion. Same point as Fig. 37, at higher magnification. × 1,100.

Fig. 39.—(Sample No. 15). Glass plate illumination. Oil immersion. Similar appearance to Fig. 38. The inclusions are of extremely small dimensions. × 1,100.

Fig. 40.—(Sample No. 15). Glass plate illumination. Oil immersion. Small transparent inclusion of vitreous silicate in process of joining a large inclusion. × 1,650.

Plate XLIV.

Fig. 41.—(Sample No. 16). Glass plate illumination. Very large inclusion of silicate of iron and manganese, found in the crop-end of an ingot, and exhibiting a fine precipitate of small globules of silica. × 330.

Fig. 42.—(Sample No. 17). Glass plate illumination. Rosettes of silica in a translucent inclusion of silicate of iron and manganese. Note the apparently transparent outline of one of the silica rosettes, which has been sectioned at the front only. × 660.

Fig. 43.—(Sample No. 17). Illumination by glass plate and by a narrow eccentric pencil of light. The same point as Fig. 42. The "cauliflower" shape of the rosettes, which are obliquely illuminated, is clearly distinguishable. × 660.

Fig. 44.—(Sample No. 15). Glass plate illumination. Silica rosettes on the vitreous ground-mass of a large inclusion of silicate of iron and manganese. × 660.

Fig. 45.—(Sample No. 15). Glass plate illumination. The same point as Fig. 44, after etching with alkaline picrate of soda. The ground-mass alone is attacked, and the silica rosettes appear in relief. × 660.

Plate XLV.

Fig. 46.—(Sample No. 16). Glass plate illumination. Small inclusion of rosette silicate. The rosettes assume the form of roughly segmented globules. × 660.

Fig. 47.—(Sample No. 15). Prism illumination. Coarse rosette type inclusion, in the same specimen as Fig. 42. After etching with alkaline picrate of soda. The silica rosettes have assumed a more markedly crystalline appearance. × 165.

Fig. 48.—(Sample No. 18). Glass plate illumination. Juxtaposition in the same field, in the region of the crop-end of an ingot, of a large, finely precipitated inclusion of silicate, a medium-sized vitreous inclusion, and a smaller "rosette" inclusion. × 165.

Fig. 49.—(Sample No. 18). Conical illumination. Same point as previous illustration, showing the relative transparency of the three types of inclusions. The true colour is brownish-yellow. × 165.

Fig. 50.—(Sample No. 19). Glass plate illumination, dry objective. Rosette silicate, showing the eutectic type of crystallisation of the ground-mass. Under a dry objective, the rosettes have a greater reflecting power than the ground-mass and are darker. × 990.

FIG. 51.—(Sample No. 19). Glass plate illumination. Oil immersion. Same inclusion as Fig. 50, at approximately the same power, but observed with an oil-immersion objective. As the refractive index of silica closely approaches that of cedar-wood oil, reflection on the plane of polishing is greatly diminished and the silica rosettes are not nearly so visible. On the other hand, the ground-mass eutectic is much clearer owing to the wider numerical aperture of the objective used. × 1,000.

PLATE XLVI.

FIG. 52.—(Sample No. 19). Prism illumination. Inclusion of silicate of iron and manganese in a rolled steel. The rosettes of deformed silica will be noted. × 330.

FIG. 53.—(Sample No. 20). Glass plate illumination. Large inclusion of basic silicate of iron in a wrought iron (Gallo-Roman iron). Light grey dendrites of ferrous oxide (or protoxide of iron and manganese) on a finely crystalline ground-mass of silicate. × 330.

FIG. 54.—(Sample No. 20). Illumination by polarised light with crossed nicol prisms. The same inclusion as Fig. 53. The protoxide is opaque and greenish-yellow. The ground-mass seems to be crystallised into lamellæ or needles. × 330.

FIG. 55.—(Sample No. 21). Glass plate illumination. Oil immersion. Complex inclusion, comprising a dark reddish-brown crystal of (Fe, Mn)O and globules of (Mn, Fe)S on a eutectic background. × 1,100.

FIG. 56.—(Sample No. 9). Glass plate illumination. Oil immersion. Typical "duplex" inclusion. Globule of vitreous silicate containing separate vesicles of sulphide. × 2,200.

PLATE XLVII.

FIG. 57.—(Sample No. 7). Glass plate illumination. Inclusion with two constituents. Globules of MnO on a background of manganese silicate. × 330.

FIG. 58.—(Sample No. 7).—Conical illumination. Same inclusion as Fig. 57. The dark-green oxide is darker than the light-green ground-mass. × 330.

FIG. 59.—(Sample No. 7) Glass plate illumination. Same inclusion as Fig. 57, after etching with dilute chromic acid. Only the background constituent is attacked. × 330.

FIG. 60.—(Sample No. 7). Glass plate illumination. Oil immersion. Same inclusion as Fig. 57, observed at a higher magnification. The ground-mass has a eutectic appearance. Certain globules of oxide are surrounded by a dark corona. They are cut through above the average middle plane, and the dark corona corresponds with the apparent contour below the polished surface observed through the translucent ground-mass. Other globules of oxide, on the other hand are cut through below the upper plane, and exhibit bright green reflections due to the light being reflected on the inner wall. × 1,100.

FIG. 61.—(Sample No. 7). Illuminated by glass plate and narrow beam of light. Point similar to Fig. 60. The ground-mass consists of a field of small and very bright green globules in a vitreous matrix. × 1,100.

FIG. 62.—(Sample No. 22). Glass plate illumination. Oil immersion. Inclusion with several constituents, formed by globules of sulphide on a lamellar eutectic ground-mass. × 1,650.

FIG. 63.—(Sample No. 22). Polarised light, illumination with crossed nicol prisms. The same inclusion as Fig. 62. The sulphides appear opaque and the ground constituent transparent, bright yellow to yellowish-brown in colour. × 1,650.

FIG. 64.—(Sample No. 23).—Glass plate illumination. Large inclusion exhibiting dendrites of manganese sulphide on a ground-mass of silicate, of lamellar eutectic appearance. × 330.

FIG. 65.—(Sample No. 23). Conical illumination. The same inclusion as Fig. 64. The sulphide dendrites are true dark-green in colour, the ground-mass being brownish-green. × 330.

PLATE XLVIII.

FIG. 66.—(Sample No. 23). Glass plate illumination. Oil immersion. Same inclusion as Fig. 64 at higher magnification. The eutectic appearance of the ground-mass will be noted. × 1,100.

FIG. 67.—(Sample No. 24). Prism illumination. "Duplex" inclusion of same type as Fig. 56, after rolling. The silicate surrounding the globules of sulphide is more drawn out than the latter. × 500.

FIG. 68.—(Sample No. 25). Glass plate illumination. Dense clusters of alumina particles in a steel as cast. × 165.

FIG. 69.—(Sample No. 25). Glass plate illumination. Oil immersion. Same sample as Fig. 59. Transparent grains of alumina. × 1,100.

FIG. 70.—(Sample No. 26). Glass plate illumination. Alumina grain of variable shape in a rolled steel. × 500.

FIG. 71.—(Sample No. 26). Glass plate illumination. Clusters of alumina elongated by rolling. × 275.

PLATE XLIX.

FIG. 72.—(Sample No. 27). Glass plate illumination. Oil immersion. Inclusions of vitreous alumina in a specimen of steel as cast. × 1,100.

FIG. 73.—(Sample No. 28). Glass plate illumination. Oil immersion. Alumina inclusions in the form of small crystals. × 1,100.

FIGS. 74 & 75.—(Sample No. 29). Glass plate illumination. Oil immersion. Inclusions of alumina crystallised into small rods. × 1,100.

FIG. 76.—(Sample No. 28). Glass plate illumination. Oil immersion. Inclusion of iron sulphide containing a small amount of (Mn, Fe)S and very small grains of alumina. × 1,100.

FIG. 77.—(Sample No. 30). Conical illumination. Large amber-yellow inclusion of silicate of iron and manganese probably containing a little aluminium in solution. × 330.

FIG. 78.—(Sample No. 30). Prism illumination. Large inclusion close to previous one and of similar character, exhibiting a precipitation commencing with small crystals of silico-aluminate of iron and manganese. × 330.

FIG. 79.—(Sample No. 30). Illumination by glass plate and narrow centred beam of light. Inclusion of silicate of iron and manganese, exhibiting an abundant precipitation of well defined crystals of silico-aluminate of iron and manganese. Focused on the background of the inclusion. × 500.

PLATE L.

FIG. 80.—(Sample No. 30). Prism illumination. Very large complex inclusion found in the vicinity of the crop-end (discard) of an ingot. Crystals of silico-aluminate of iron and manganese on a ground-mass crystallised into needles or lamellæ. × 330.

FIG. 81.—(Sample No. 30). Conical illumination. Same inclusion as Fig. 80. The crystals are reddish-brown and the ground-mass pale yellow. The acicular structure will be very clearly distinguished. × 330.

Plate LI.

Fig. 82.—(Sample No. 30). Glass plate illumination. Oil immersion. Same inclusion as Fig. 80, at higher power. Two twinned crystals of silico-aluminate. The ground-mass is formed by transparent, pale yellow dendrites. × 1,100.

Fig. 83.—(Sample No. 30). Prism illumination. Vitreous inclusion of silicate containing a large idiomorphic crystal of silico-aluminate of iron and manganese. × 330.

Fig. 84.—(Sample No. 30). Conical illumination. Same inclusion as Fig. 83. The plane of polishing passes approximately through the top left-hand corner of the crystal. × 330.

Fig. 85.—(Sample No. 31). Conical illumination by mirror. Partial view at higher power of one of the extremities of the inclusion depicted in Fig. 86. The crystals are in relief on the ground-mass. This photomicrograph frequently gives the impression of reversed relief. The true impression can be restored by varying, for example, the direction in which the photomicrograph is observed. The light is incident in the direction shown by the arrow. × 165.

Plate LII.

Fig. 86.—(Sample No. 31). Prism illumination. Very large inclusion of the type shown in Fig. 80, but after rolling. Deformed and broken crystals of silico-aluminate of iron and manganese in an apparently vitreous matrix. × 65.

Plate LIII.

Fig. 87.—(Sample No. 31). Prism illumination. Partial view at higher power of part of the field of Fig. 85. Dendrites of manganese-iron sulphide, of light greyish-blue colour, are visible between the silico-aluminate crystals in the ground-mass. × 500.

Fig. 88.—(Sample No. 31). Glass plate illumination. Same point as Fig. 87, after etching with chromic acid. The sulphide dendrites are completely dissolved. The ground-mass is slightly attacked. × 500.

Fig. 89.—(Sample No. 31). Glass plate illumination. Same point as Fig. 88, after etching with alkaline sodium picrate. The ground-mass is heavily attacked. × 500.

Fig. 90.—(Sample No. 31). Glass plate illumination. Same point as Fig. 89, after etching with hydrofluoric acid. The ground-mass is entirely dissolved, and the crystals of silico-aluminate detached or partially corroded. Note the bright reflections which are due to total reflection from the back faces of the crystals. × 500.

Plate LIV.

Fig. 91.—(Sample No. 31). Illumination by glass plate with a narrow beam of light. Oil immersion. High-power examination of the sulphide dendrites shown by arrow in Fig. 87. Focused on the front branches. × 1,100.

Fig. 92.—(Sample No. 31). Illumination by glass plate with a narrow beam. Oil immersion. Same point as Fig. 91, but focused on the side branches of the dendrites. The true colour is bright green. × 1,100.

Fig. 93.—(Sample No. 32). Glass plate illumination. Inclusion of the same type as Fig. 86, of smaller dimensions, and containing comparatively few crystals of silico-aluminate. The latter are not very visible with glass plate illumination. × 275.

Fig. 94.—(Sample No. 32). Glass plate illumination. Same point as Fig. 93, after etching with fused potash. Only the vitreous ground-mass is blackened. × 250.

FIG. 95.—(Sample No. 33). Prism illumination. Inclusion of the same character as the ones in the previous illustrations, almost exclusively composed of crystals of silico-aluminate, there being very little vitreous binder. × 165.

FIG. 96.—(Sample No. 33). Glass plate illumination. Lower part of the inclusion as Fig. 95, after etching with alkaline sodium picrate. The crystals of silico-aluminate suffered very little deformation or breakage, and have retained their original idiomorphic form. × 500.

FIG. 97.—(Sample No. 33). Glass plate illumination. Same point as Fig. 96 after etching with hydrofluoric acid. Same remarks as for Fig. 90. × 500.

PLATE LV.

FIG. 98.—(Sample No. 26). Glass plate illumination. String of crystals of silico-aluminate of iron and manganese in rolled metal. × 500.

FIG. 99.—(Sample No. 26). Glass plate illumination. Oil immersion. Juxtaposition in the field of view of colourless inclusions of alumina (left) and silico-aluminate (right) × 1,100.

FIG. 100.—(Sample No. 30). Glass plate illumination. Oil immersion. Translucent inclusion which seems to have two constituents — globules of sulphide (Mn, Fe) S and vitreous silicate. × 1,100.

FIG. 101.—(Sample No. 30). Glass plate illumination. Oil immersion. Same inclusion as Fig. 100, but microscope focused on the bottom of the inclusion. A crystal of silico-aluminate will be noticed near a globule of sulphide. × 1,100.

FIG. 102.—(Sample No. 30). Glass plate illumination. Oil immersion. Two crystals of silico-aluminate combined with a globule of sulphide. × 4,400.

FIG. 103.—(Sample No. 31). Glass plate illumination. Oil immersion. Mixed inclusion of the same type as Fig. 100, after rolling. Note, encased in the very malleable vitreous silicate, a globule of sulphide that has not been deformed very much (*a*), and a crystal of unchanged silico-aluminate (*b*). × 2,200.

PLATE LVI.

FIG. 104.—(Sample No. 34). Glass plate illumination. Globular inclusion of vitreous silicate containing a few coarse grains of alumina. × 250.

FIG. 105.—(Sample No. 34). Conical illumination. Same inclusion as Fig. 104., The ground-mass silicate is not very transparent compared with the alumina, which also shows up in marked relief. × 250.

FIG. 106.—(Sample No. 35). Glass plate illumination. Very large silicate inclusion containing numerous grains of alumina. A few metallic globules. × 660.

FIG. 107.—(Sample No. 36). Glass plate illumination. Oil immersion. Large inclusion formed by the juxtaposition of grains of alumina with very little binder. × 1,100.

FIG. 108.—(Sample No. 25). Glass plate illumination. Oil immersion. Inclusion as Fig. 107, surrounded by numerous clusters of alumina. × 165.

FIG. 109.—(Sample No. 37). Glass plate illumination. Inclusions of alumina, and crystalline inclusion probably formed by a silicate of alumina. × 660.

FIG. 110.—(Sample No. 37). Glass plate illumination. Same point as Fig. 109. The inclusions are colourless. × 660.

CORRESPONDENCE.

Mr. L. E. BENSON (Manchester) wrote that Professor Portevin and Mr. Castro had added a further contribution of merit to the growing mass of data regarding the composition and identification of inclusions in iron and steel. There was still, however, a great need for any test which afforded a ready method of discrimination between different kinds or types of inclusions. In that connection the accompanying illustrations might be of interest to Professor Portevin and Mr. Castro or others working on the same lines.

Fig. A (Plate LVIa.) showed inclusions as seen on a longitudinal section of molybdenum steel containing approximately 0·3 per cent. of carbon and 0·5 per cent. of molybdenum. The section was cut from a specimen that had extended 25 per cent. in length in 3½ hr. under steady tension at 550° C. It would be noticed that the inclusions were of two kinds, one of which had broken into blocks in a non-ductile manner. Fig. B depicted a longitudinal section of a sample of free-cutting mild steel bar which appeared to have been finished by cold-rolling or cold-drawing. The inclusions were again of two distinct kinds, and again, one kind, the darker in colour, was broken into blocks, whilst the other inclusions were intact.

The analyses of the steels were as follows :

	C.%	Si.%	Mn.%	S.%	P.%	Mo.%
Molybdenum steel .	0·29	0·17	0·57	0·016	0·042	0·61
Mild steel . .	0·16	0·01	0·71	0·205	0·101	..

In both the above cases the inclusions appeared to have behaved in a ductile manner during hot-rolling. Further, it had been proved that the rupturing of the darker inclusions in the molybdenum steel occurred during the test at 550° C., and it was presumed that in the mild steel specimen the rupturing occurred during cold-working.

The possibility of the ductility of inclusions at normal or moderately high temperatures affording at least a useful check for identification purposes appeared to be worth further investigation, particularly as any method of testing likely to be devised on these lines would be quicker and less laborious than most existing chemical methods.

Mr. J. H. WHITELEY (Consett) wrote that the authors had, he believed, contributed a most useful paper on the many kinds of inclusions to be found in steels. It was probably correct to say that, given a good chemical analysis, the quality of a steel depended very largely upon the types, sizes and quantities of the inclusions

present. Their identification was often extremely difficult, and the information given in the paper should be very helpful. Two methods which he had found of value in certain cases were, however, not mentioned by the authors. They were the acid gelatine reagent, and heat-tinting.[1] The former, when applied to the silico-alumina inclusions, for example, should readily reveal the presence of any sulphides. The identification of the silico-aluminate of iron and manganese constituent was particularly valuable, since this inclusion was to be seen almost invariably in steels which had been deoxidised with alsimin. No reference had been made to a rather common inclusion in ordinary steels, consisting of strings of broken silicates, described many years ago by Stead, who showed that they consisted of a silicate of iron and manganese and were rather basic in character. The writer's own experience confirmed this, for these broken strings appeared to be confined to rather over-oxidised steel containing about 0·02 per cent. of silicon, yet they were not very plastic and appeared to be an exception to the authors' statement that the moderately basic silicates were much more malleable than vitreous silica and sulphides (pp. 255 and 258). In studying the origin of the inclusion content of a steel one of the methods used by the writer was to examine sections from the forged bath sample taken on tapping an open-hearth furnace and compare it with sections of the finished steel. Time and again he had found that the quantity to be seen in the latter was much the greater. In the bath sample the inclusions were usually extremely minute, requiring a high power for their observation, whereas the finished steel might contain quantities easily visible under low magnification. In such cases any one of several of these in a single area, if ground fine, would make sufficient for the content of many such areas in the bath sample. As far as he had observed, there appeared to be frequently a marked growth in the bulk quantity of inclusions during casting and teeming. It would be of much interest to have the authors' observations and views on this point.

AUTHORS' REPLY.

The AUTHORS were much obliged to Mr. Benson for the micrographs of inclusions which he had been kind enough to submit, as well as for the interesting suggestions which he had put forward regarding the relative malleability of inclusions at normal and reduced hot-working temperatures. Regarding the micrographs, they offered the following comments: Fig. A, according to the authors, represented basic silicates that were

[1] *Journal of the Iron and Steel Institute*, 1920, No. II. p. 145.

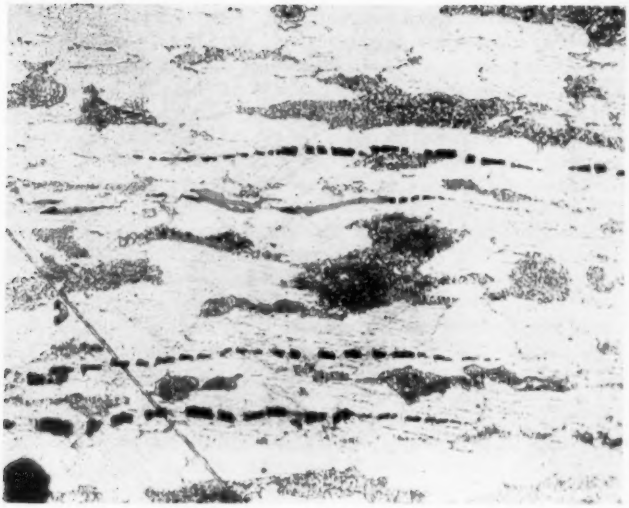

FIG. A.—Inclusions in Molybdenum Steel, stretched 25 per cent. × 400.

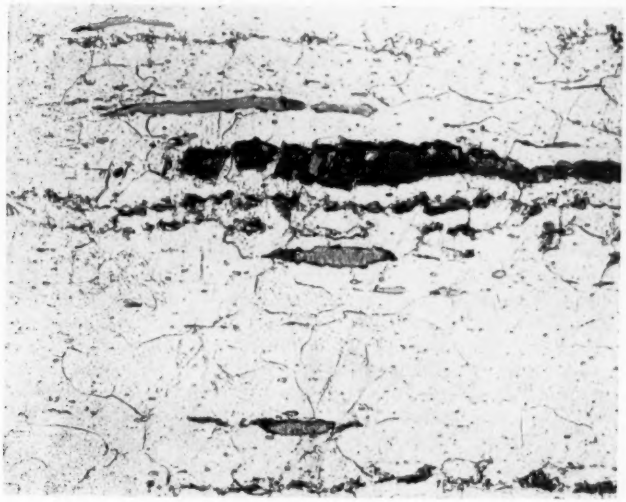

FIG. B.—Inclusions in Free Cutting Steel. × 400.
(Micrographs reduced to four-fifths linear in reproduction.)
(*See* Benson's Communication.)

Portevin and Castro Correspondence

To face p. 276.

very malleable at normal hot-working temperatures (dark-coloured strings). As Mr. Benson supposed, these silicates were decidedly less malleable at lower temperatures and could break up into strings of fragments ; the hot-working having been carried out slowly, the metal surrounding the inclusions had been sufficiently plastic to fill the spaces between the fragments. On the contrary, sulphides—which in the present case should consist almost entirely of MnS, having regard to the manganese content of the metal—were malleable right down to room temperature and were not broken. The second string of inclusions in Fig. A, commencing from the top, had the appearance of a " duplex " silicate-sulphide inclusion after deformation. Fig. B, on the other hand, showed that the dark inclusion which was not malleable in the cold had broken without the metal being able to penetrate into the spaces with the same facility. It was difficult

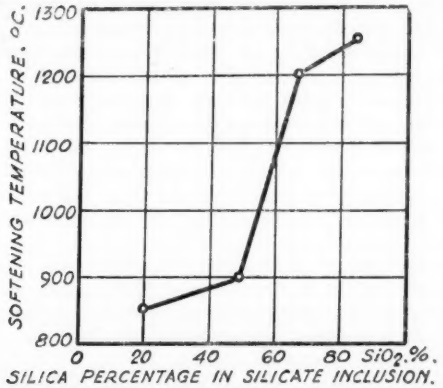

FIG. C.—Relation between the Softening Points of Silicate Inclusions from Four Steels and their Silica Contents (Schnell and Scheil).

to determine the exact nature of this inclusion solely from the appearance of the micrograph, but, nevertheless, it was probable that, having regard to the analysis of the steel, the phase (Fe, Mn)O was present. Examination under conical and polarised illumination would then show the reddish or red-yellowish colour characteristic of the non-metallic material. The clear phase consisted of manganese sulphide which was malleable in the cold.

The modification of inclusions by deformation of the metal at various temperatures had already been studied, principally by Scheil and his collaborators, in recent publications.[1] That

[1] R. Schnell and E. Scheil, *Mitteilungen der Vereinigten Stahlwerke, Dortmund*, 1934, vol. 4, p. 235.
E. Scheil, *Zeitschrift für Metallkunde*, 1935, vol. 27, pp. 202-204.

study was intended in particular to show the relation between the degree of deformation of the metal and that of the inclusions at different temperatures, but it had equally shown up very clearly the influence of the silica content (see Fig. C), as determined analytically, on the softening points of the silicates. In this connection Scheil said : " This variation is so large that it would be worth while to use the softening temperature of the oxidised inclusions in a metal as a criterion of the nature of those inclusions."

Thus Mr. Benson's suggestion was perfectly logical and had already found a partial application. It must not be forgotten, however, that such a method of valuation would be accurate only in the clearly restricted case of vitreous silicates of iron and of manganese ; in fact, the addition of another element, such as aluminium or chromium, which could modify the composition of these silicates in a manner unpredictable except by means of analysis, would introduce an element of uncertainty by probably causing the softening point of these phases to vary ; further, this method appeared to the authors to be applicable only in the case of silicates containing a single vitreous phase, the slightest commencement of crystallisation diminishing the plasticity. Finally, if it was a case of distinguishing two different plastic phases, such as oxides or silicates and sulphides, their micrographic appearances and their optical properties were sufficiently different in general for this to be done satisfactorily by microscopic examination.

The determination of the softening point of vitreous non-metallic inclusions by hot-working at various temperatures was thus an ingenious method of expressing a new physical property of these substances and of obtaining in certain cases more precise data regarding their chemical nature.

The authors thanked Mr. Whiteley for his contribution to the discussion on their work. They were acquainted with his interesting researches on the determination of inclusions, principally those of sulphides, by means of an acid gelatinous solution of tartar emetic. The reason why no mention had been made of this method in their paper was that the authors were more particularly concerned to define still more closely the external appearance and the physical and optical properties of inclusions than their behaviour towards various chemical reagents. In general, sulphides, when isolated, were sufficiently discernible and optically recognisable for it to be unnecessary to have recourse to an etching method, but just in the case of silico-aluminous inclusions—the aluminium, as had been said, probably throwing the sulphides out of solution from the silicates—the sulphides might be enclosed in a mass of oxides in such a way as to be scarcely visible ; then it was that Whiteley's method could show

them up very clearly by means of the yellow rings of antimony sulphide which formed round these minute grains.

With regard to the broken silicates reported by Mr. Whiteley, the authors would refer to certain remarks which they had made in reply to Mr. Benson. The ductility of a silicate inclusion was essentially a function of the temperature at which its deformation had taken place. It was thus very probable that the strings of basic silicates, referred to above, occurred in steels of which the hot-working had been carried out at a relatively low temperature; conversely, by hot-working at abnormally high temperatures it should be possible to draw out very acid silicates without breaking them. The relative characteristics of malleability of duplex silicate-sulphide inclusions mentioned by the authors were valid only for normal rolling or forging temperatures, and were susceptible to complete reversal at lower temperatures, as was shown clearly by Fig. A put forward by Mr. Benson.

Conversely, these interesting observations could be used to obtain data regarding the thermal conditions of mechanical treatments undergone by the metal; having determined the nature of the inclusions from their optical and chemical characteristics, deductions could be drawn from their condition and morphological characteristics as to the temperature zones within which the deformations of the metal were effected.

The authors had also observed the fact that spoon samples often appear to be cleaner than the corresponding finished metal. Estimations of the total oxygen carried out on these two materials had shown that their oxygen contents were comparable, or even that that of the finished metal was lower; the phenomenon arose, in fact, from the impossibility of determining precisely the quantity of inclusions present in steel by micrographical means. The recent researches of a large number of metallurgists, particularly Körber and his collaborators, had shown that during the cooling of a more or less oxidised liquid metal containing silicon and manganese, inclusions precipitated continuously right up to the moment of the solidification of the metal. Spoon samples were of very small bulk and solidified rapidly, and the precipitated inclusions were, in fact, of extremely small dimensions; in particular, if such a sample were examined under the highest possible magnifications, the existence of inclusions of which the fineness was of the same order as the resolving power of the objective could be observed. At the same magnifications examination under conical illumination revealed still more. It was thus practically certain that the metal thus rapidly cooled contained a very large quantity of inclusions so fine that they escaped microscopical observation. Perhaps in this case it was feasible that a little oxygen might remain in solution in the

metal, but this point was still very obscure and had not yet been clearly proved.

On the contrary, metal solidifying in large ingot moulds cooled slowly, the inclusions formed round less numerous nuclei and in particular had a tendency to coalesce strongly, as was shown in Figs. 37 to 40. In this case the metal, when examined under conical illumination at very high magnifications, appeared to be free from very small inclusions. Thus the subjective microscopic valuation of the quality of the non-metallic material included in a sample of steel could lead to sensible errors of interpretation and quantitative analytical determination only was valid in this case.

This observation was important in the sense that it furnished yet another argument against the use of micrographic examination as a method of quantitative evaluation of inclusions for use as a means of checking the quality of steel, to the defects of which the authors had already drawn attention.[1]

[1] R. Castro and A. Portevin, *Revue de Métallurgie, Mémoires*, 1932, vol. 29, p. 558.

THE BEHAVIOUR OF MILD STEEL UNDER PROLONGED STRESS AT 300° C.

PART II.—EXPERIMENTS ON CONCENTRATED STRESS IN NOTCHED AND DRILLED SPECIMENS.[1]

By C. H. M. JENKINS, D.Sc., A.R.S.M.

(National Physical Laboratory)

Résumé.

The tests on specimens subjected to a stress locally concentrated at drilled holes and machined notches have now been in operation for a period of five years. Although considerable local deformation has occurred at the points of stress concentration, no cracking has developed, indicating that localised stress, in the absence of a chemical accelerating agent, is insufficient to cause cracking in such parts as boiler plates and straps.

The experiments described in the present paper form a continuation of a previous set of experiments on the behaviour of mild steel in regard to the possibility of cracking under the action of prolonged stress at 300° C. The late Dr. Rosenhain and Professor Hanson[2] showed that samples of a mild steel (material *A* of the present paper), when subjected to various forms of heat treatment, were capable of withstanding severe stresses up to two-thirds of the normal breaking stress, for prolonged periods at 300° C. They claimed, as a result of many tests lasting five years, that ultimate fracture was unlikely, but showed that the specimens which had been subjected to stresses between one-third and two-thirds of the normal breaking stress had become considerably hardened. The rate of straining was low and was applied uniformly throughout the cross-section of the test-piece. They suggested that there was a possibility that localised stress such as that due to a notch might give different results.

The present paper, which summarises an investigation of the Metallurgy Research Board of the Department of Scientific and Industrial Research, is devoted to the study of localised stress produced by the use of notches and drilled holes in similar strip specimens. The suggestion of the previous authors in regard to the possible effect of localised stress has not been proved to be correct. In addition to the low-carbon and low-manganese

[1] Communication from the National Physical Laboratory received June 26, 1935.

[2] W. Rosenhain and D. Hanson, "The Behaviour of Mild Steel under Prolonged Stress at 300° C." *Journal of the Iron and Steel Institute*, 1927, No. II. p. 117.

steel (*A*) used in the previous work, specimens of a boiler-plate steel (*B*) have also been tested. This plate, $\frac{9}{16}$ in. thick, had been withdrawn from service in a boiler after failure by intercrystalline weakness between rivet holes. The specimens were cut parallel to the length of the boiler, that is, at right angles to the curvature of the plate. The strength of this material *B*, tested on British Standard type *C* bars, 0·358 in. in diam., was as follows :

Yield stress. Tons per sq. in. . . .		22·5	21·2
Ultimate stress. Tons per sq. in. . .		33·1	32·6
Elongation (gauge length, $4\sqrt{A}$).	% . .	34·0	34·0
Elongation (gauge length, $8\sqrt{A}$).	% . .	25·0	25·0
Reduction of area at fracture.	% . .	59·0	58·0

The compositions of these steels were as follows :

	Mild Steel *A*. %	Boiler Plate *B*. %
Carbon	0·11	0·23
Silicon	Trace	0·02
Sulphur	0·075	0·033
Phosphorus	0·050	0·066
Manganese	0·395	0·72

The effect of difference of microstructure was determined on each material after three forms of heat treatment, which are given below (no change in microstructure occurred during the test) :

Material *A*.	Material *B*.
Normalised.	In service condition.
Spheroidised at 650° C.	Normalised.
Spheroidised by slowly cooling through critical range.	Spheroidised at 650° C.

The minimum load placed on any specimen was somewhat below one-half the yield point at room temperature, and the maximum load in three cases just exceeded the yield point. The results of the tests on materials *A* and *B* are recorded in Tables I. and II., respectively.

Two forms of test specimen were used in the present tests ; in one type the stress was concentrated by a drilled hole, in the other by notches machined in both edges of the specimen.

(a) *Notched Specimens.*—Each specimen, which on the parallel length was 0·25 in. wide and 0·02 in. thick, contained three notched areas, and each specimen can be regarded as three separate test-pieces. The form of specimen is shown in Fig. 1. Each notched area was formed by machining a 45° notch on the edge of the specimen so as to leave a residual width of 0·10 in. The radius at the root of each of the six notches was 0·25 mm.

(0·01 in. approximately). The gauge length was 0·50 in., which was marked midway between the notches, and the whole of the movement of each notched area was, therefore, measured. It is apparent that the portion of the specimen which *deformed* under the load was very short and was not likely to exceed 1–2 mm.

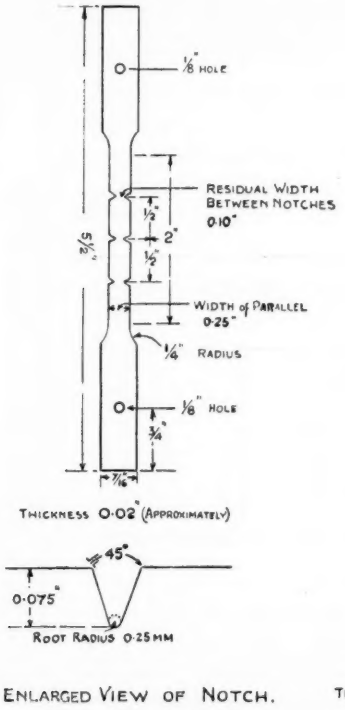

FIG. 1.—Notched Specimen.

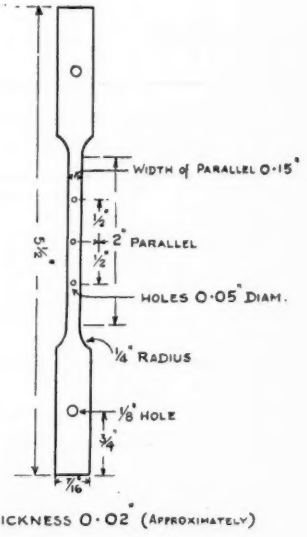

FIG. 2.—Drilled Specimen.

(b) *Drilled Specimens.*—Each drilled specimen, similarly to the notched specimens, presented three opportunities for studying the effect of a concentrated load. Drilled holes, 0·05 in. in diam., were made through a strip specimen, 0·15 in. wide by 0·02 in. thick, in the positions shown in Fig. 2. In order to introduce the effect of cold-work, each drilled hole was impressed on both sides of the strip by a 2-mm. hardened steel ball under a 50-kg. load for ½ min.

TABLE I.—*Test Results on Mild Steel* A.

Tensile tests at 15° C.; prolonged tests at 300° C.

Tensile Test on Unnotched or Undrilled Strip Specimens.			Specimen Mark.	< = Notched. o = Drilled.	Previous Heat Treatment.	Average Load on each Notched, &c., Zone. Tons per sq. in.	Total Elongation on 3 Notches or Holes after 1,826 Days. In. × 10^{-3}.	Largest Max. Load on Individual Notched, &c., Zone. Tons per sq. in.	Elongation on Notched, &c., Zone with Largest Max. Load. In. × 10^{-3}.	Reduction of Area at Notched, &c., Zone with Largest Max. Load. Per cent.
Yield Point. Tons per sq. in.	Max. Load. Tons per sq. in.	Percentage Elongation. 2-in. Gauge Length.								
			A6CS	<	Heated to 900° C. Spheroidised by slowly cooling from 720°–650° C. in 6 hr.	7·6	8·7	7·7	3·1	N.D.
11·8	18·3	29	*A7CS*	<		11·56	14·5	11·6	6·7	4·0
12·4	18·5	29	*A8CS*	o		7·48	5·1	7·55	2·4	N.D.
			A9CS	o		11·0	31·0	11·3	10·2	4·0
			*A5N**	<	Normalised from 900° C.	15·53	50·0	16·1	18·5	18·0
15·2	22·0	38	*A6N*	<		11·56	12·6	12·0	4·3	N.D.
15·2	21·4	33	*A8N*	o		12·06	10·6	12·2	3·9	N.D.
			A9N	o		13·9	83·1	14·0	28·0	26·0
			A6S	<	Spheroidised for 4 days at 650° C.	7·8	5·9	7·9	2·4	N.D.
11·3	18·5	33	*A7S**	<		11·9	19·3	12·0	5·9	6·0
11·7	19·7	32	*A8S*	o		7·8	6·7	8·0	1·1	N.D.
			*A9S**	o		11·33	28·7	11·7	11·4	4·9

* Specimens *A5N*, *A7S* and *A9S* were stressed above the yield point of the material, which varied with the microstructure.
Accuracy of measurement of gauge length = ± 0·01 mm. = ± 0·0004 in.
N.D. = not determined.

TABLE II.—*Test Results on Mild Steel* B.

Tensile tests at 15° C.; prolonged tests at 300° C.

Tensile Test on Unnotched or Undrilled Machined Strip Specimens.			Specimen Mark.	< = Notched. o = Drilled.	Previous Heat Treatment.	Average Load on each Notched, &c., Zone. Tons per sq. in.	Total Elongation on 3 Notches or Holes after 1,826 Days. In. × 10^{-3}.	Largest Max. Load on Individual Notched, &c., Zone. Tons per sq. in.	Elongation on Notched, &c., Zone with Largest Max. Load. In. × 10^{-3}.	Reduction of Area at Notched, &c., Zone with Largest Max. Load. Per cent.
Yield Point, Tons per sq. in.	Max. Load. Tons per sq. in.	Percentage Elongation. 2-in. Gauge Length.								
			B3S	<		7·15	0·4	7·4	0·4	N.D.
15·2	24·1	22	*B4S*	<	Spheroidised for 4	11·5	Nil	12·4	0·4	N.D.
14·9	23·2	15	*B5S*	o	days at 650° C.	7·86	0·4	9·9	0·8	N.D.
			B6S	o		11·0	5·1	11·1	2·0	3·0
			B11N	<		11·26	Nil	11·4	Nil	N.D.
19·8	32·6	20	*B12N*	<	Normalised from	14·55	4·7	15·2	2·4	6·0
21·6	32·5	11	*B13N*	o	900° C.	11·03	1·6	11·1	1·2	2·0
			B14N	o		15·66	4·7	16·0	2·4	2·0
			B17R	<		10·73	0·4	11·8	0·4	N.D.
23·5	33·2	14	*B18R*	<	As received, "ser-	14·23	Nil	14·5	0·4	N.D.
22·3	32·4	19	*B19R*	o	vice condition."	12·56	Nil	13·9	0·4	N.D.
			B20R	o		14·23	Nil	14·9	0·4	N.D.

All specimens were stressed below the yield point, which varied with the microstructure.
Accuracy of measurement of gauge length = ± 0·01 mm. = ± 0·0004 in.
N.D. = not determined.

A comparison of the notched and the drilled specimens shows that the curvature at the point of minimum cross-section was 2·5 times as great in the notched as in the drilled specimens. The

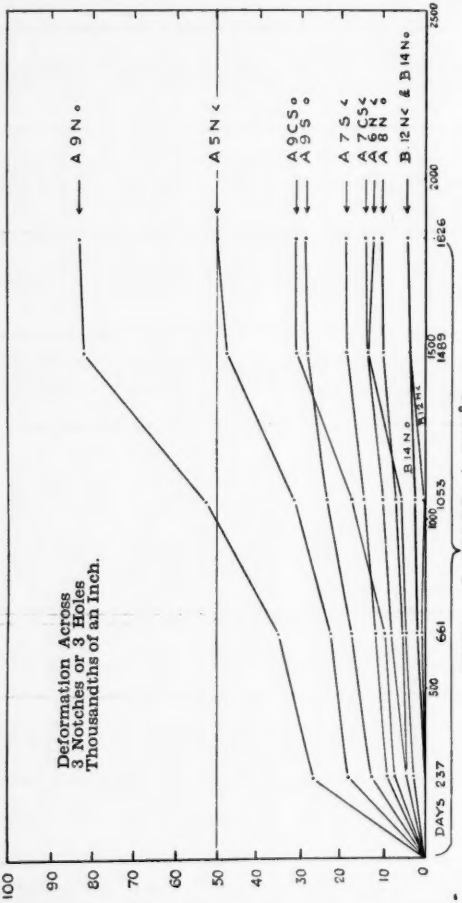

FIG. 3.—Test Results.

results recorded in Tables I. and II. show a much greater extension in the case of drilled than in notched specimens, and the results suggest that the concentration of stress is not only dependent on the shape of the specimen but varies from one material to another and from one type of heat treatment to another.

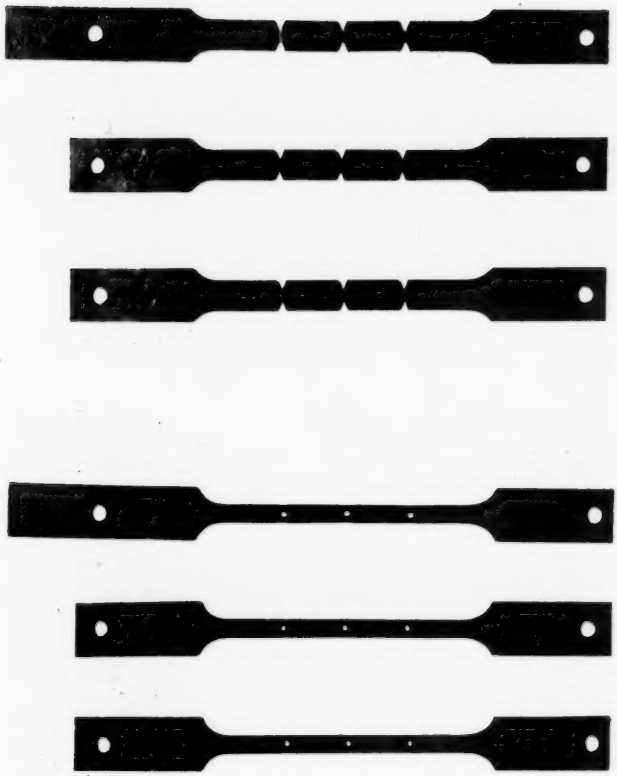

FIG. 4.—Specimens of Material *A*, after five years at 300° C.

[*Jenkins.*
[*To face p.* 286

The degree of deformation around the notches or around the holes is indicated in Fig. 4 (Plate LVII.), and the extension of certain specimens during the course of testing is shown in Fig. 3. The twenty-four specimens were withdrawn from the oven heated to 300° C. for measurement after periods of 237, 661, 1,053, 1,489 and 1,826 days, and a confirmatory set of specimens was tested for a further 237 days, together with about six new specimens of material *A*. This further set of specimens confirmed the experimental conditions and the initial rapid extension of the material and the small extension in the fifth year of testing.

Conclusions.

Although considerable deformation has taken place around the holes and notches inserted in the specimen, failure by cracking has not been produced under a stress similar to that of the yield point. It would appear that concentration of stress, cold-work and a temperature of 300° C. are not sufficient to bring about failure, and that tests with chemical accelerants are desirable.

The author wishes to thank Mr. H. E. Bennett, formerly of the Metallurgy Department, and latterly Mr. G. C. H. Jenkins, of the same Department, for assistance during the course of this investigation.

CORRESPONDENCE.

Mr. C. W. HIRST (Sheffield) wrote that on first reading this paper and arriving at the point where the test-pieces were described, he formed the very definite opinion that, owing to the presumably greater localised stress at the bottom of the notches, the notched pieces would elongate more than the drilled pieces. The figures presented in the later part of the paper showed the opposite result, and he wondered whether the result would have been quite the same had the width of the parallel portions of all the test-pieces been the same, namely, 0·15 in., with a notch depth of 0·025 in. or one-half of the hole diameter.

He mentioned this because, in certain tests on Izod impact pieces of constant section behind the notch, but with varying notch depth, the variation in the energy required to break the pieces was much greater than could be explained by the variation in the impact property of the material.

As the material which was stretched lay a little to either side of the notch centre, it appeared possible that the extra width of the notched piece might exert a buttressing effect on the material which was undergoing extension.

He would also ask Dr. Jenkins if he had found any measurable reduction of area on the parallel portions of the drilled specimens.

AUTHOR'S REPLY.

The AUTHOR thanked Mr. Hirst for his contribution and agreed that the concentration of stress was probably greater in the notched specimens than in the drilled specimens, and that, therefore, it would be expected that deformation would be greater in the notched material. There was against that effect a greater length of effective material in the drilled than in the notched specimens. To bring about a true comparison, tests on specimens of similar width, with the same radius of curvature in the notches and in the drilled holes should be made. The notched specimens would require to be shaped with semi-circular notches. The object of the tests on notched specimens was to provide a high concentration of stress at the root of the notch. The author was grateful that Mr. Hirst had pointed out the buttressing effect of wider parallel-sided specimens.

The author regretted that he had not any direct measurements of the lengthening of the parallel portions of the test-pieces of drilled specimens, but careful note had been made of the amount of lengthening that occurred at the drilled holes. For example,

in specimen *A9N* the test-piece had lengthened 2·11 mm., whereas the three holes had elongated 0·94 mm. The stress on the parallel portion had been 10·5 tons per sq. in. In a second specimen, *AL9*, three holes had elongated 0·28 mm. and the whole specimen stretched 0·73 mm. under a stress of 7·8 tons per sq. in. on the parallel portion. The stress at the holes was recorded in the original table. There was, of course, appreciable stretch in the immediate region of the holes, and undoubtedly stretching had occurred on the actual parallel portion of the test-piece clearly away from the holes. The five-year tests on parallel-sided specimens described by Rosenhain and Hanson[1] indicated the degree of lengthening of similar material by a comparison of the cross-sectional areas of undeformed test-pieces.

[1] *Journal of the Iron and Steel Institute*, 1927, No. II. p. 117.

SIXTH REPORT ON THE HETEROGENEITY OF STEEL INGOTS.

By a JOINT COMMITTEE of the Iron and Steel Institute and the British Iron and Steel Federation to the Iron and Steel Industrial Research Council.

This Report is published as Special Report No. 9; it was presented for discussion at the Autumn General Meeting of the Iron and Steel Institute, held in Manchester, on September 17, 1935. As arrangements have been made for it to be discussed further at a number of Joint Meetings with local Societies, to be held during the winter months, the publication of the discussion and correspondence, and the Committee's reply, has been deferred until it is possible to deal with them comprehensively. They will be published with the proceedings of the Annual Meeting, 1936.

WASTE-HEAT BOILERS IN OPEN-HEARTH PRACTICE.

SECOND REPORT OF THE OPEN-HEARTH COMMITTEE, BEING A COMMITTEE OF THE IRON AND STEEL INDUSTRIAL RESEARCH COUNCIL.

This Report is published as Special Report No. 10; it was presented for discussion at the Autumn General Meeting of the Iron and Steel Institute, held at Manchester, on September 18, 1935. As arrangements have been made for it to be discussed further at a number of Joint Meetings with local Societies, to be held during the winter months, the publication of the discussion and correspondence, and the Committee's reply, has been deferred until it is possible to deal with them comprehensively. They will be published with the proceedings of the Annual Meeting, 1936.

VISITS AND EXCURSIONS AT THE MANCHESTER MEETING.

THE Autumn Meeting of the Institute was held in Manchester during the week commencing Monday, September 16, 1935. A large Reception Committee, representing the Iron, Steel and Engineering Industries of Manchester and Lancashire and the Iron and Steel Industries of the West Coast of Cumberland, had been formed, and of this, the Right Hon. The Earl of Crawford and Balcarres, P.C., K.T., was the President, Mr. J. E. James the Chairman, Alderman F. J. West the Hon. Treasurer and Mr. H. D. Lloyd the Hon. Secretary. The details of the preparations were put into the hands of the following Committees:

EXECUTIVE COMMITTEE.

G. E. Bailey.	H. E. D. Mabbott.
G. Bentham.	Col. J. G. Needham, D.S.O., O.B.E.
C. W. Ellis.	K. Treherne Thomas.
J. E. James.	F. J. West.

PUBLICATIONS COMMITTEE.

Professor F. C. Thompson, *Editor*.	Charles Reynolds.
J. Sinclair Kerr.	Geo. Douglass.

TRANSPORT COMMITTEE.

C. W. Ellis. H. G. Humphreys. R. Stuart Pilcher.

LADIES' COMMITTEE.

Mrs. G. E. Bailey.	Mrs. H. D. Lloyd.
Mrs. J. E. James.	Lady Rylands.
Mrs. V. Z. de Ferranti.	Lady Simons.
Mrs. P. Holloway.	Mrs. K. Treherne Thomas.
Mrs. J. Sinclair Kerr.	Mrs. F. J. West.

The thanks of the Council and members are due to the above-mentioned ladies and gentlemen, and particularly to Lord Crawford, the President, Mr. J. E. James, the Chairman, Alderman F. J. West, the Hon. Treasurer, and Mr. H. D. Lloyd, the Hon. Secretary, for the excellent manner in which the arrangements were planned and carried out, and to Mr. R. Crichton (of Moss Bay) and Major A. Hibbert (of Millom) for their care and attention in organising the excursion to Cumberland. Each member was presented with a most interesting book entitled "The Iron, Steel and Engineering Industries of Manchester and District," which, as its compilers intended, formed a most useful souvenir of the meeting; besides an article on "Historical Manchester," by Professor G. W. Daniels, M.A., M.Com., and another on "The Iron, Steel and Engineering Industries of Manchester and District," by Mr. John Jewkes, M.Com., it contained brief descriptions of the works of the Companies which were to be visited by the members during the meeting as well as those of other firms which would be of interest to the visitors.

By the kindness of Principal B. Mouat Jones and the College Authorities, rooms in the Manchester College of Technology were made available for the Meeting. Several departments of the College of Technology were thrown open for inspection by the members on the evening of Monday, September 16, when they called to collect their programmes, &c., and guides were in waiting to show them round.

The first session of the general meeting was held in the Great Hall of the College of Technology on Tuesday, September 17, at 9.30 A.M. At the opening of the proceedings, The Rt. Hon. The Earl of Crawford and Balcarres, P.C., K.T. (President of the Reception Committee), and the Lord Mayor of Manchester (Alderman Samuel Woollam, J.P.) attended to give the Institute a very cordial welcome. An account of the proceedings will be found in the first few pages of the present volume, but occasion may be taken here to express the Council's endorsement of Sir Harold Carpenter's thanks to the Lord Mayor and Lord Crawford for the warmth of their welcome and the interest which they were taking in the success of the meeting.

Later in the morning, while the members were engaged in the discussion of the papers presented, the ladies were taken by motor coach to visit the showrooms of the Tootal Broadhurst Lee Co., Ltd., and also Rylands Library, the New Free Library and Chetham's Library.

At 12.30 P.M. the members and Ladies reassembled at the Midland Hotel, where they were entertained to lunch at the invitation of the Reception Committee under the Chairmanship of Lord Crawford, the President of the Committee.

The Loyal Toast having been duly honoured, the CHAIRMAN (Lord Crawford) said: I am going to give you only one toast, but one which I am sure you will accept with acclamation. It is a toast expressing our thanks to the Principal and to the Trustees of the College of Technology for their kind hospitality to the Institute. I wish to associate it, if I may, with the name of Mr. Mouat Jones, the Principal. We in Lancashire are particularly glad to see him back at his post after a very long and serious illness. (*Applause.*)

At the meeting this morning I felt as if I were in an examination hall, and after about twenty minutes of the discussion I said to myself "What on earth can heterogeneity mean?" (*Laughter.*) I therefore took up the Sixth Report of the Committee, which was placed in front of me, and opened it casually at page 13, where I found the following question: "Small particles, too small to obey Stokes' law, settle to the lower third of the mould." (*Laughter.*) This was followed by the word "Why?" (*Renewed laughter.*) I felt that that was directly addressed to myself. Is it owing to the modesty of the small particles, or to their defiance? Anyhow, the small particles must have great courage to disobey Stokes' law! (*Laughter.*) But what is Stokes saying about it? (*Laughter.*) And if laws, whether of Stokes or of others, be so easily defied, how can you gentlemen ever expect to solve more than what Lord Rutherford says you have only solved, that is to say one per cent. of the problem of heterogeneity?

I have been haunted by this question (*laughter*), and I have been wondering how to solve it. It was not proper on my part to ask Sir Harold Carpenter during the progress of this morning's research, and at this luncheon he has shown no inclination to embark on these subjects at all; so that, if I am pressed, I have come to the conclusion that the only thing to do is what is commonly done in the examination hall by Mr. Mouat Jones's students, and that is to evade the reply (*laughter*), to answer by example or by analogy. In this great centre of Lancashire commerce—and may I say of English commerce too—I will give you an example of an evasive reply. A youth asked his father for a definition of commercial morality. "It is," said his father, "a very difficult thing to define"—like Stokes' law (*laughter*)—"but I will give you an example. If I lend a man £100 and the man pays me back £120 by accident, immediately a point of commercial etiquette arises: ought I or ought I not to tell my partner Ikey?" (*Loud laughter.*)

Taking all the circumstances into account, I have come to the conclusion that I shall abandon the attempt to probe into the niceties of Stokes' law. I am going to lie low; I am going to imitate the small particles; I am going to get down to the bottom third of the mould (*laughter*), and there I shall repose, there I

shall settle, and so, if the answer to the question is to be given, it must be given by Mr. Mouat Jones, after you have drunk his health with cordiality and enthusiasm. (*Prolonged applause.*)

Mr. B. Mouat Jones (Principal of the Manchester College of Technology), who was cheered on rising to reply, said : Feeling as I do like a very small particle, unable to settle, I should like to do as Lord Crawford has done and evade a reply. I do not see why I should be inflicted on you in this way ; after all, you have never done me any harm. (*Laughter.*) I can only suppose that it is some form of sanctions.

The College of Technology feels itself very highly honoured in having been selected as the place for your deliberations. Looking through the programme of your activities, I tried to find out what these deliberations would be about, and, after a long search among the excursions and the social events of various kinds, I discovered them. (*Laughter.*) When I found them I was reminded of a certain wartime incident which happened on an English steamship. There was a conjuror on board who was entertaining the passengers, and he had just said " Ladies and gentlemen, I will now make this parrot disappear " when the ship was blown sky-high by a German torpedo. As they picked up the survivors, they found a somewhat bedraggled-looking bird strutting up and down a floating plank and muttering to itself " Damned clever ! " (*Loud laughter.*)

I should like to thank Lord Crawford very cordially for the way in which he has proposed this toast. I will throw a few sidelights on the work of the College. A technical college has a very great variety of subjects of instruction, but not nearly enough to meet with the demands of the public, who are always seeking enlightment on every conceivable subject. I have been asked to consider classes in such things as piano-tuning, beauty culture, lion-taming (*laughter*), chemistry for chicken farmers, saw-doctoring, embalming, monumental masonry and a host of other similar cheerful occupations ; and the other day I received a letter addressed to the " College of Knowledge, England," which was correctly delivered without delay (*laughter*), and in which the writer asked me to provide for him a course of instruction as an insecticide. (*Laughter.*)

Even in the subjects with which we do deal, it is not always possible to meet the immediate needs of every individual. The other day I had a letter from a parent asking me whether his son could give up Scripture and take Commercial French, as it would be more useful to him hereafter. (*Loud laughter.*) We run a course for Health Visitors, and teach the scientific feeding of young children. At a recent examination the candidates were asked to give three reasons why natural feeding of infants is

superior to artificial feeding; and one young lady, confusing science with common sense (*laughter*), replied: "First, because the temperature is right; secondly, because the cream cannot be taken off it; and thirdly, because the cat can't get at it." (*Loud laughter.*) That gives you a fair indication of the way in which we train our students to grasp the essentials of a problem and to deal with problems when they arise with practical and commonsense methods. (*Laughter.*)

I could tell you a good deal more about the work of the College (*laughter*), but I must remember that you, like the metal with which you deal, are capable of suffering fatigue. I have only to wish your Institute a very successful, enjoyable and improving visit to Manchester. I feel it would be kind to remind those of you who are going on the Manchester Ship Canal to-morrow of the only poem that has been written about any Manchester institution—

> "Mary had a little lamb,
> Its fleece was white as snow.
> It went into the Ship Canal—
> And now look at the damned thing!" (*Laughter.*)

Sir HAROLD CARPENTER, D.Sc., F.R.S. (President of the Institute) said: I am sure you would not wish to separate without my expressing in a sentence our thanks to the Reception Committee for the delightful hospitality that we have enjoyed. (*Applause.*)

The proceedings then terminated.

In the afternoon (Tuesday) arrangements had been made for the members to visit one or other of the following works:

> Metropolitan Vickers Electrical Co., Ltd., Trafford Park.
> Messrs. Mather and Platt, Ltd., Park Works, Newton Heath.
> Messrs. Taylor Bros. & Co., Ltd., Trafford Park.
> Renold and Coventry Chain Co., Ltd., Renold Works, Didsbury.
> Messrs. Rylands Bros., Ltd., Warrington.

The members were conveyed in motor coaches to the works, where they were received by the Principals and members of the staffs, who acted as guides and conducted the parties round the works. The visitors were entertained to tea before returning to their hotels. While the members were thus engaged, the Ladies were taken to visit the new housing scheme of the Manchester Corporation at Wythenshawe Park; they also were provided with refreshments before the return journey.

On Tuesday evening, at 8.0 P.M., the Lord Mayor (Alderman Samuel Woollam, J.P.) and the Corporation of the City of Manchester offered the members and their Ladies a Civic Reception in the Town Hall. Later in the evening this was followed by a Concert and Dancing.

On Wednesday morning, September 18, the second session for the discussion of papers was held in the Students' Common Room at the College of Technology. At its conclusion the members were taken by motor coaches to Salford Docks, where they embarked on steamers, kindly provided by the Manchester Ship Canal Co., for the Irlam Works of the Lancashire Steel Corporation, Ltd. On arrival, the guests took lunch, at the invitation of the Corporation, under the Chairmanship of Mr. John E. James (Chairman of the Company). Then followed visits to all the departments of the works, which were followed by refreshments before the party left on the return journey by motor coach.

In the meantime, the Ladies spent the day on an excursion to Chatsworth, where a visit to Chatsworth House and grounds was paid by permission of His Grace the Duke of Devonshire, K.G. (Past-President of the Institute). The outward journey was made *via* Stockport and Chapel-en-le-Frith. Lunch was taken at the Palace Hotel, Buxton, and later the party returned over the Cat and Fiddle, *via* Macclesfield and Knutsford. At the latter place they were entertained to tea at Ollerton Grange at the invitation of Mr. and Mrs. John E. James.

At 7.30 P.M. on the same evening (Wednesday), the members and their Ladies were entertained at a Banquet and Dance at the invitation of the Reception Committee. At the Banquet The Rt. Hon. The Earl of Crawford and Balcarres, P.C., K.T., occupied the Chair ; the guests numbered over four hundred.

At the end of dinner, after the Toast of " His Majesty the King, Duke of Lancaster " had been duly honoured, the following Toasts were proposed :

" The City, Port and Trade of Manchester."

Mr. FRED J. WEST, C.B.E., M.Inst.C.E., M.I.Mech.E. (Chairman of the Manchester Ship Canal Company) said : I have the honour to submit the toast of the City, Port and Trade of Manchester. However inadequately I put forward the claims of this toast, I feel sure it will be received by you with the sincerity and enthusiasm which it deserves. The toast is of a very comprehensive nature ; it affords scope for a great variety of expression, and indeed it is difficult to fix on what might be termed the correct starting point. If I start with the City of Manchester—and seeing that we have with us this evening the chief citizen, the Lord Mayor of Manchester, who will respond to this toast—I feel

sure that the members of the Iron and Steel Institute would wish me to say to the Lord Mayor how very much they appreciated his welcome to the City yesterday at the opening session and enjoyed the reception yesterday evening at the Town Hall; further, they would all like to record their intense satisfaction that the Lord Mayor from a civic point of view has done everything that lies in his power to make the Autumn Meeting of the Institute in Manchester a success.

The Lord Mayor typifies that form of local government which has placed this country in the forefront of the nations of the world with regard to our public activity and our health and social services. I have endeavoured, in my travels to Australia, New Zealand, the United States, Canada and South Africa, to make a study of local government matters, and I have come to the very definite conclusion that our form of local government, by comparison with that of other countries, has a great deal to be said for it and in many respects is superior to others. I would further say that Manchester has no need to be ashamed of the position which it occupies amongst the large provincial cities with regard to its contribution to the form of local government which prevails in this country.

The City of Manchester has much of which to be proud. Her University, her College of Technology and her other educational facilities are, as you were told by Principal Mouat Jones yesterday, exceedingly practical. Her social and philanthropic institutions, to say nothing of Manchester's commercial standing and the business acumen of her leading citizens, have placed the city in a leading position of which we are not unreasonably proud.

With regard to the second portion of my toast, the Port, I think many of you will be surprised to hear that Manchester is the fourth largest port in this country, and that we bring the largest freight-carrying steamers over our 32 miles of waterway direct to Manchester from all parts of the world. We handle, with our imports and exports, something like six million tons per annum, and we have created, in the forty-one years for which the waterway and port have been operating, something like a business world's record, with far-reaching consequences to the commerce and industry of the great areas surrounding Manchester.

It is of interest to find that in the early days of the construction of the canal, in 1887, the Iron and Steel Institute paid a visit to the waterway and inspected the progress of the work, and it is still more interesting that the first Chairman of the Ship Canal, Mr. Daniel Adamson, who held that post from 1885 to 1887, was President of the Iron and Steel Institute at the time of that visit.

During our trip on the canal to-day, I observed Lord Crawford, our Chairman this evening, making a very concentrated study of the water in the canal. (*Laughter.*) I imagine he was admiring

the beautiful bluish tint of the water, and I saw how he could apply Stokes' law (*laughter*) to explode a fallacy which exists with regard to the behaviour of the water of the Manchester Ship Canal. Had he been proposing this toast, I believe he would have further expounded the law to which he referred yesterday by the use of one of the terms in the formula, the coefficient of viscosity. (*Laughter.*) He would thus have exploded the fallacy, which is a libel on the canal, that it is impossible for a man to be drowned in it, the suggestion being that he would be poisoned first. (*Loud laughter.*)

With regard to the Trade of Manchester, although the city is known to many as Cottonopolis, and although it is correct to say that it is the centre of the cotton industry, fortunately Manchester has a great variety of trades, and it may be a surprise to many here to learn that the engineering trades, in which most of us here are interested, employ more workpeople than the cotton industry itself in the area surrounding Manchester, which is to-day one of the most important engineering centres in the country.

We have in Manchester an organisation known as the Manchester Engineering Council which has been very successful in promoting happy relations between the employers and the workpeople. We claim that in Manchester there is a closer and better understanding between them than exists in many other centres. We have freedom of speech at our meetings, and we do talk very freely to each other.

Time will not permit me to say anything further with regard to the trade of Manchester, and I will conclude by saying on your behalf to the Lord Mayor and citizens of Manchester, on this, the last function in connection with your Autumn Meeting here, that you earnestly hope that the City and Port of Manchester and the Trade of Manchester and the great area surrounding it will continue to develop and to prosper in a manner worthy of Manchester's distinguished history and past achievements. (*Applause.*)

The toast was received with enthusiasm.

The Rt. Hon. THE LORD MAYOR OF MANCHESTER (Alderman Samuel Woollam, J.P.), who was received with applause on rising to respond, said: I cannot consider the way in which your Chairman has arranged this toast list as a very kindly act on his part. When I promised him to do everything I could to make your meeting a success if you honoured us with a visit to Manchester, I thought I had made a friend of him (*laughter*); but now I find that he has selected as the proposer of this toast one who has not only proposed the toast but who has also very ably

responded to it. (*Laughter.*) Moreover, when I came here to-night I thought of reciting a little poem to you about Mary and her little lamb, and now I find you have heard it already ! (*Laughter.*)

The proposer of this toast is the Chairman of the Manchester Ship Canal, and our Corporation has derived great benefit from the support which it gave in the 'eighties to enable that canal to be completed. The Corporation invested over £6,000,000 of capital in the Ship Canal Company, and I believe in 1960 the money which it borrowed for that purpose, something like £5,000,000, will have been paid back, and the ratepayers of the City will then enjoy the advantages which will accrue to them from the interest paid on that money.

I remember that when I came to Manchester in 1878 I was told that it was no use going to Manchester because it was down and out, and there was no likelihood of a raw youth from the country getting employment. I took the risk and found employment, although unfortunately or fortunately for me, in 1908 I joined the ranks of the unemployed, and without the dole !

Mr. West, in proposing this toast, has done us the honour of associating with it the Corporation of the City of Manchester. When I tell you that we have no fewer than 33,000 persons employed by the Corporation, that we have a rateable value of over £6,000,000, and that three of our trading committees have a turnover of over £3,000,000, it will enable you to realise that those who are privileged to serve as members of the Corporation have some responsibility to bear in connection with its work. We know that our rates are high, but I would ask those of you who are privileged to live in Manchester to think for a moment of the facilities provided for those rates. I think I can say that there is no city or town in the United Kingdom that gives better value to the citizens for the rates that they have to pay, and so long as the citizens feel they are getting value for their rates it must be a source of great satisfaction to the city's representatives. My friend Mr. Mouat Jones and those associated with the College of Technology and our junior technical schools have over 3,000 students at the present time training for the engineering and chemical industries. I hope that some of them may at a later date become useful members of this Institute.

My friend the Chairman of the Chamber of Commerce would be better able than I am to tell you something about the trade of the city, because during his year of office he has been concerned with many committees and sub-committees dealing with this difficult problem of how to bring more trade not only to Manchester but to Lancashire. I am quite sure, from what I know of him and his colleagues that no stone is left unturned to enable industry to come to Manchester and to Lancashire and provide employment for our people.

I thank you, Sir, for the cordiality with which you have proposed this toast, and you ladies and gentlemen, for the way in which you have received it. I can say, on behalf of my colleagues on the Council, as well as on behalf of the Lady Mayoress and myself, that it has been a real pleasure to do anything we could to make your meeting a success. (*Applause.*)

"*The Iron and Steel Institute.*"

The CHAIRMAN (The Rt. Hon. The Earl of Crawford and Balcarres, P.C., K.T.), who was warmly applauded on rising to speak, said: It is a great happiness as well as a great honour to propose the toast of the Institute. We cordially welcome them to our midst, for it is long since we have had the honour of receiving them; but, as the Lord Mayor has just pointed out, we in Lancashire are shy and retiring people (*laughter*) who do not always do full justice to our virtues or to our enterprise, and we are therefore at times in danger of being overlooked, although our merits are incalculable. (*Laughter.*) Occasionally, however, we do penetrate the great world beyond our borders and learn lessons in many things, some of which leave us a little surprised, and perhaps nonplussed as well.

Two citizens of Manchester recently got as far as London, and on their return could be persuaded to say nothing but "There's queer chaps in Lunnun." (*Laughter.*) On being pressed for an explanation, it was discovered that they had enjoyed their visit very much, and on the last morning they determined to go to one of the great shops and purchase mementoes of the occasion; the man was to buy some gold studs or links and his wife a muff. The man had the greatest difficulty in explaining in the shop what he wanted, the reason being that he said he wanted to buy some "fancy tackle." They tried to sell him every kind of fish-hook (*laughter*), but ultimately he reached the jewellery department and was shown trays of gold rings and sleeve links and studs and all kinds of fancy tackle, and the assistant, trying to make him buy something, said it was eighteen carat. "Ating carrot?" said the man, "Nay, Ah'm chewing bacca!" (*Loud laughter.*) His wife said she wanted a "mooff," and the polished young Londoner inquired "What fur, madam?" "What fur?" she said, "Wha, to kip me 'ands warm!" (*Laughter.*) "Ay, but there's queer chaps in Lunnun!" (*Laughter.*)

Well, we are very grateful to London and very grateful to the Institute, but during the last twenty-four hours I have felt particular need of saying how grateful I feel to Yorkshire—a very surprising thing to say! (*Laughter.*) Mr. West has just referred to Stokes' law, and I made a reference to Stokes' law yesterday. Since then no fewer than five members of the Institute have explained to me, in very nice language and with good

humour, what Stokes' law is. I think I should perhaps say that in their definitions there is a very strong element of heterogeneity. Nothing they have said to me has convinced me that the small particles that ought to have obeyed Stokes' law were anything but wise in disregarding it. I certainly should not like to have to give you myself a definition of what Stokes' law really is; my mind is in great confusion on the subject and, like Mr. Mouat Jones's scholars, I shall give you a wrong answer to the right question. They teach divinity in the School of Technology at Manchester, and one of his youths was asked what he could say about Dan and Beersheba. "Oh," he replied, "they are just a married couple in the Old Testament, like Sodom and Gomorrah." (*Laughter.*)

That brings me by a very easy transition to the toast of the Iron and Steel Institute. (*Loud laughter.*) On your behalf, I wish all success and all prosperity to the Institute. We here well recognise the important task that it fulfils and the great obligations it carries in the direction and control of the technical side of these leading industries. The importance of the Institute grows day by day as the obligations placed upon the trade and the industry are increased by the State.

I have the honour, in proposing this toast to you, to couple with it the name of the Institute's distinguished President, Sir Harold Carpenter. In wishing success and prosperity to the Institute, we equally wish it to the great industry upon which the security of our country fundamentally depends. (*Applause.*)

The toast was cordially received.

Sir HAROLD CARPENTER, D.Sc., F.R.S. (President of the Iron and Steel Institute), who was received with acclamation on rising to speak, said in reply: At an early stage in the dinner, Lord Crawford consulted me as to the line he might take for his speech. I said "I am going to say exactly what I like, and I hope you will do the same." I should like to congratulate him and to thank him for the delightful words he has used in presenting this toast.

As I mentioned yesterday, Manchester has been a good friend to this Institute, especially in its early days when it needed friends; and in our first thirty years of existence we came here no fewer than three times. When my wife and I visited the city last Monday evening, one of the members of the Institute said to me "This is coming home for you, isn't it?" That was a very true saying, and perhaps truer than he realised. I am very grateful to him for saying that, because it constitutes the peg on which I am going to hang the few words I have to say.

I came to this city in 1898, thirty-seven years ago, at the very beginning of my career, and, although I am a poor southerner,

I chose Manchester as the city from which to begin that career. I have never regretted the choice, and I am convinced that I could not have made a better one. I hope there are some in this room who remember some of the old Owen's College professors of those days. I came at the time of Osbourne, Reynolds, Schuster, Dixon and Bargate, and there was also a man in Manchester then whose name I am sure many will remember, Sir Henry Roscoe. Lord Crawford is Chancellor at this moment of one of the most distinguished and prosperous Universities in this country, but it has not always been so. I will tell you a story that I heard Sir Henry Roscoe tell just about that time. To-day we returned from the Ship Canal up Key Street, which is where the College used to be. One evening Sir Henry was smoking a cigar outside, preparatory to giving his evening lecture, when a tramp passed and said "Is this the night shelter?" He replied "Well, it isn't at present, but if you call again in a month it may be!" (*Laughter.*) As a matter of fact, the number of students then was just about thirty.

After three years in Manchester, I went to London to the National Physical Laboratory. We started there with a staff of five, and three of them were Manchester men; we used to regard the National Physical Laboratory as a little bit of Manchester. In 1906 the University did me the honour of inviting me to return here as its first Professor of Metallurgy. That is twenty-nine years ago, and I want to tell you a little about the men who were my first demonstrator, my first student and my first laboratory assistant. They are all of them in this room, and their wives also.

My demonstrator began life in the foundry of the old Lancashire and Yorkshire Railway at Horwich. He had come to me at the National Physical Laboratory on special work, and when I came here he told me he would like to come with me. Some of my Manchester friends were generous enough to subscribe sufficient money to get him here for three years. I asked him whether he would take the risk, and he said he would. He therefore came for three years, and I explained that in that time he could get what he had not got then, a degree. Presently he married, and then a child came. You cannot live on a demonstrator's salary, and so he went off to Bolckow, Vaughan & Co. and to Dorman, Long. When I went to London, that man succeeded me here. Fifteen years ago he went to Swansea as the first Professor of Metallurgy there, and since then he has become Principal of the College. He has been Vice-Chancellor of the University and he is a Fellow of the Royal Society. You have doubtless guessed that I am referring to Principal Edwards. (*Applause.*)

My second example is my first student. I am not sure that he came from Lancashire; I believe he came from Cheshire. He

has done two very remarkable things. He succeeded in obtaining a post in a Scottish University (*laughter*), and at the present time he is showing them in Yorkshire how to do things. He has exhibited, as the Principal of the University of Sheffield mentioned when we were there two years ago, a degree of indefatigability which is the envy of his colleagues. I am referring now to Professor Andrew. (*Applause.*)

I am not going to make any comparisons, but I am not sure that I am not most interested in my third example, whom I will frankly call a laboratory boy. This boy had the invaluable quality of making a piece of apparatus work. (*Laughter.*) As to that, I am reminded of a definition which was once given of the difference between a theoretical man and a practical man. The theoretical man, it was said, is a man who explains how a machine works; the practical man is the man who explains why it does not. (*Laughter.*) Now, my laboratory boy had that second quality in a very high degree; nothing went right when he was away from the laboratory, and we were very unhappy. He ran us all, and, in addition to that, in a little room—a sort of Black Hole of Calcutta—he ran a chess tournament. After that he went off into the foundry industry, and to-day he is a member of the Council of the Institute of British Foundrymen, he is chairman of their chief technical committee, and he is a consultant for foundry firms in the district of Birmingham. I refer to John Gardom. (*Applause.*)

I have quoted these instances to you for my personal pleasure and as an example to you of the degree of initiative and enterprise which you say exists in Lancashire, and I entirely endorse that view. The particular conditions which obtain this evening will never hold again, and that is why I have ventured to strike this personal note and to tell you something about these three men. If I may, I should like to ask you to join with me in congratulating Professor and Mrs. Andrew, Principal and Mrs. Edwards and Mr. and Mrs. Gardom on the great success which these men have achieved in life. (*Prolonged applause.*)

Coming to the present day, we always try, when we visit a city, to obtain papers from local members, and I wish to allude particularly to the very interesting and valuable paper presented this morning by Professor Thompson. (*Applause.*) Professor Thompson is carrying out a very special line of work on wire-drawing, and in fact on the drawing of metals generally, which I am sure is of deep interest and importance to this City. I should also like to pay my tribute to the valuable work which Professor Bragg is carrying out in the Department of Physics on the structure of alloys.

Before I sit down I want to do what I am sure the members of the Iron and Steel Institute would like me to attempt to do,

namely, to express our thanks to the Reception Committee for the really wonderful time we have had. (*Applause.*) To some extent Mr. West has anticipated me in the references which I was proposing to make to the Lord Mayor, but I must be allowed to say, as President of the Institute, that we appreciated very highly the speech which the Lord Mayor made yesterday, and which I for one will never forget; it was a very remarkable and indeed a perfect speech. Coming to to-day, I learn from my wife of the very delightful trip which our ladies were taken to the Peak District, and we think especially there of Mrs. James, who was so kind as to entertain them in a manner which delighted them very much. (*Cheers.*) The local Committee arranged a very interesting trip for the male members, and we went from the School of Technology to Irlam by way of the Ship Canal. There has been a number of references to that piece of water already, but I propose to draw attention to an aspect of the matter which has not been commented on so far. Reference has been made in particular to the colour of the water, but I was much more appreciative of, and I may add affected by, the emanations from it. (*Laughter.*) Mr. West was very desirous of telling me that the particular odours we encountered to-day had not been met with for six years. I am quite prepared to believe that, but I do maintain that there is evidence for the fourth line of the poem which the Principal of the College of Technology gave us yesterday (*laughter*), and I think the matter should receive further attention.

I am going to couple with this toast the name of Mr. James. (*Applause.*) I met Mr. James only for the first time in May, and I was put next to him at the Institute Dinner in London because it was then known that I was to be President and he was to be Chairman of the Local Reception Committee. Perhaps I can express my opinion of him best by saying that after that very brief meeting I had not the slightest anxiety about the success of this visit; there was a quiet competence about him which means everything, and I knew that we were going to have a grand time. We have had it, and I am going to ask the members of the Institute to rise and give Mr. James musical honours. (*Applause.*)

The toast was drunk with musical honours.

Mr. JOHN E. JAMES (Chairman of the Reception Committee) said in reply: I should like to thank you for the way in which you have paid tribute to the little effort that we have made to try to give this meeting the success that it deserves. When it was impressed on me that it was about time that the Institute came to Manchester again, I felt considerable diffidence about taking on the responsibility of organising the visit. I should like to say how deeply grateful I am, as a comparative stranger,

for the very many kindnesses that I have received on all sides from the leaders of industry in Manchester. Perhaps the outstanding and shining light has been the Lord Mayor. Had it not been for that encouragement, I do not think that it would have been possible for me to carry out this programme.

Throughout the meeting we have been treated to the most humorous side of Manchester; I feel there can be no doubt that the music hall artists of the future will come from the School of Technology. (*Laughter.*) It has been a very great pleasure to arrange this visit of the Institute, and I hope they will not be so long in returning to Manchester as they were last time.

Before I sit down I must pay my tribute to the President of the Reception Committee. Lord Crawford has helped more than he would ever know or admit. I feel that had it not been for that great grace and charm which is his natural possession much that we have done as a Reception Committee would have been impossible. (*Applause.*)

It has been one of the greatest pleasures and proudest moments of my life to think that I have been able at long last to contribute something to the Iron and Steel Institute, with all its great traditions and with all that it has done for the industry. This is the first time that I have taken any really important part in its activities. Now that I have set my foot on the threshhold and introduced to Manchester what I consider the cream of the world so far as industry is concerned, I feel sure that they in return will realise that in Manchester they have discovered people whom they never knew existed before. (*Laughter.*) Perhaps they have also discovered other things that they never knew existed before.

In conclusion, I beg to thank you for your very great kindness. (*Applause.*)

After the Banquet, the floor was cleared and dancing was continued for the rest of the evening.

It should be added that, besides the works officially visited as mentioned above, other Companies had expressed their willingness to receive members of the Institute during the afternoons of Tuesday and Wednesday, September 17 and 18; the following is a list of these Companies:

Churchill Machine Tool Co., Ltd., Broadheath, Altrincham.
Messrs. Craven Bros. (M/C), Ltd., Vauxhall Works, Reddish, Stockport.
Messrs. Ferranti, Ltd., Hollingwood, Manchester.
National Gas and Oil Engine Co., Ltd., Ashton-under-Lyne.
Messrs. Geo. Richards & Co., Ltd., Broadheath, Altrincham.

Messrs. Henry Simon, Ltd., Cheadle Heath, Stockport.
West's Gas Improvement Co., Ltd., Albion Ironworks, Miles Platting, Manchester.

The whole of Thursday, September 19, was devoted to an excursion into the Lake District. The party travelled by special train from Victoria Station at 8.15 A.M. Windermere was reached before 11 A.M., and motor coaches carried them to the Keswick Hotel for lunch. After lunch the members travelled on by the train (which had come round to Keswick), to Workington, where they visited the Moss Bay Works of the United Steel Companies Ltd. Taking tea on the train, they then proceeded to the works of the Millom and Askam Hæmatite Iron Co., Ltd., at Millom, where special arrangements allowed the train to run right into the Company's works. After a short tour of inspection, followed by refreshments, the members again entered the train, to make the return to Manchester, *via* Ulverston. At both works the members were received by representatives of the Boards of Directors and officials of the Companies.

Meanwhile the Ladies had been conveyed by motor coaches through Threlkeld, Troutbeck, Aera Force, Ullswater, Patterdale, Kirkstone Pass, Bowness, Newby Bridge and Grenold on the way to Ulverston. Tea was taken at the Bellsfield Hotel, Bowness, and the members of the party had the option of leaving the coaches there, travelling by steamer to Lakeside, and rejoining the coaches there.

At Ulverston the Ladies met the members' special train, and the last section of the return journey to Manchester was made by that means; dinner was served on the train.

At 9.37 P.M., the party arrived back at Victoria Station, and so ended the fourth visit of the Iron and Steel Institute to the City of Manchester.

OBITUARY.

Dr. ARTHUR BRAMLEY died on July 19, 1935, at the age of fifty-six. He had to leave school early and served a full apprenticeship to the hosiery trade. However, he became an evening student of Halifax Technical College and in 1904 obtained a staff appointment there. In 1906 a national scholarship took him to the Royal College of Science, where he graduated, carried out research work, and finally became private research assistant to Professor J. C. Philip. His researches included studies of the physico-chemical properties of solutions and ionic complexes, and an investigation of the physical properties of binary liquid mixtures, for which work he was awarded the degree of D.Sc.

Dr. Bramley subsequently served on the scientific staff of the British Dyestuffs Corporation at Huddersfield, where he remained until he was appointed to take charge of the Department of Pure and Applied Science, Loughborough College, in 1918. While at Loughborough he undertook a long series of researches on the diffusion of non-metallic elements into iron and steel, and was able to show that the mechanism of this diffusion follows the normal physico-chemical course. For these researches he was awarded the Carnegie Gold Medal in 1929. An enthusiastic teacher and brilliant research worker, he had been a member of the Iron and Steel Institute since 1922, and had been responsible for numerous papers published in *The Carnegie Scholarship Memoirs* during the years 1926 to 1929.

WILLIAM ROBERT DAVIES, of St. Petroc, Park Road, Whitchurch, Cardiff, died on May 15, 1935, at the age of sixty-three. Born at Lydbrook, Forest of Dean, in 1872, he received his early training with Messrs. Richard Thomas & Co., Ltd., and his first official appointment was at the Aberdare Tinplate Works in 1895. In 1902 he became assistant works manager of the Melingriffith Company, of Cardiff, and five years later was appointed general works manager; he became a Director of the company in 1920, and later Joint Managing Director with Mr. H. Spence Thomas, from which position he retired in June 1934, when the works were acquired by Messrs. Richard Thomas & Co., Ltd. Mr. Davies was technical partner in the development of the Thomas and Davies machines, &c., owned and run by the company. These devices included the combined automatic feeding, pickling and tinning machine known as the "Melingriffith pot," as well as doubling, shearing and other machines. Mr. Davies was, until he retired, Vice-Chairman of the Welsh Plate and Sheet Manufacturers' Association and Chairman of

the Joint Industrial Council and the Joint Standing Committee Wages Board. He was a member of the Institution of Mechanical Engineers and of the South Wales Institute of Engineers; he joined the Iron and Steel Institute in 1908.

Sir JAMES ALFRED EWING, K.C.B., D.Sc., F.R.S., died at Cambridge on January 7, 1935; he was seventy-nine years old. Born at Dundee on March 27, 1855, he was educated at the High School in that city, whence he obtained the first engineering scholarship to Edinburgh University. At that time Fleeming Jenkin was working there with Kelvin on submarine telegraphy, and he suggested that Ewing should take part in this work during the vacations; after graduating, he made three voyages on cable-laying ships to Brazil and the River Plate. In 1878, on his chief's nomination, he was appointed Professor of Mechanical Engineering at Tokyo University; during his five years' stay in Japan, he took the opportunity to study seismology; he designed instruments by which the exact measurement of earthquake motion was made practicable, and these were installed in an observatory which he established in Tokyo. He also began researches which later gave birth to the molecular theory of magnetism, this accounting for the phenomenon to which he gave the name hysteresis.

In 1883 he returned to Britain and became Professor of Engineering at University College, Dundee; seven years later he followed James Stuart, the first occupant, in the Chair of Mechanism and Applied Mechanics at Cambridge. After his appointment, on the advice of John Hopkinson, considerable developments took place; a laboratory was acquired, the general principles for the education of engineers were laid down, and a Tripos was established—all this was due in no small measure to Ewing's energy.

While at Cambridge, Ewing continued his magnetic researches, paying special attention to the practical testing of the magnetic properties of iron; the results were published in a paper presented before the Institution of Civil Engineers in 1896. He also devised a magnetic curve tracer, a hysteresis tester, and a permeability bridge, while in another field he investigated the crystalline structure of metals and determined how an aggregate of crystals can exhibit plasticity through the slipping processes that occur between atomic layers.

In 1903, his connection with Cambridge was interrupted, and was not resumed for some years; when the Admiralty decided to introduce a new Scheme of Naval Education, he was first called upon to advise how best to include a large element of engineering knowledge in the training of every officer, and then was appointed Director of Naval Education. While at the

Admiralty he was a Member of the Explosives Committee and of the Ordnance Research Board. At the beginning of the Great War, he was invited to assist in the de-coding of enemy ciphers, cryptograms being a subject in which he had long been interested as an amateur ; as a result, a new department, known as Room 40, was set up.

In 1917, Sir Alfred was appointed Principal and Vice-Chancellor of the University of Edinburgh, but he continued his association with Room 40 for another year. At the end of the war, problems of University reconstruction arose, and he showed himself to be an able administrator, with resource, bold leadership and sound business capacity. During his tenure of office, thirteen new chairs were established, many new buildings were erected, and the whole institution was reorganised on modern lines. He resigned his Vice-Chancellorship in 1929 and retired to Cambridge, where he was elected an Honorary Fellow of King's College. While he was at Edinburgh, the Bridge Stress Committee was appointed by the Department of Scientific and Industrial Research, in March 1923, of which body he was Chairman ; the work of this Committee, which consisted of the experimental determination of the static and dynamic deflections of every ordinary type of bridge and of the stresses in individual members, lasted for five years, and showed the effect on the oscillation of bridges of two factors—the hammer blows of the unbalanced parts and the frictioned resistance of the spring gear of the locomotives—which had previously not been taken fully into account.

At Cambridge, Sir Alfred voluntarily devoted some of his time to other activities of the Department of Scientific and Industrial Research. As a member of the Food Investigation Board he was interested in the work of the Low-Temperature Research Station in that town, and he supervised researches at the National Physical Laboratory on thermal insulation, refrigerants and the measurement of humidity—a subject to which he had earlier paid much attention. He was also a member of the Committee on the Mechanical Testing of Timber, appointed by the Department in 1929.

Sir Alfred was the author of many papers on scientific subjects, particularly magnetism and the physics of metals. His treatise on "The steam engine and other heat engines," published in 1894, has become a classic, and has been translated into many languages. His other works include books on "The strength of materials," on "The mechanical production of cold" and on "Thermo-dynamics for engineers."

Sir Alfred was naturally the recipient of many honours. He was made a Companion of the Bath in 1907, and became Knight Commander of the same Order four years later. He was

elected a Fellow of the Royal Society in 1887, and received the Royal Medal for his researches on magnetism in 1895. He held honorary degrees in the Universities of Oxford, Cambridge, Durham and St. Andrews, and was elected an Honorary Member of the Institution of Civil Engineers in 1929, in which year also he received the Albert Medal of the Royal Society of Arts. He was a Past-President of the Royal Society of Edinburgh, and was awarded the freedom of that city and of his native town of Dundee. He had been a member of the Institution of Civil Engineers since 1891 and of the Institution of Electrical Engineers since 1893, and had served on the Council of the latter body. He joined the Institution of Mechanical Engineers in 1891, and was made an Honorary Life Member in 1932. His membership of the Iron and Steel Institute dated from 1899.

ROLAND CHARLES HARDING died at his home in Llantarnam, Monmouthshire, in September, 1935, at the age of fifty-one. He was the youngest son of the late Mr. Joshua Harding, of Stoke-on-Trent, and received his training with Messrs. Robert Heath and Sons; later he became Managing Director of the firm. In 1930 he became general manager of the Rogerstone Works of Messrs. Guest, Keen and Nettlefolds, Ltd., which position he held at his death. He joined the Iron and Steel Institute in 1912.

WILLIAM HIPKINS, of Edgbaston, Birmingham, died in April, 1935. He was probably the oldest ironmaster in the Midlands; he was in his ninety-second year, and had been retired from active business for some years. He joined the Iron and Steel Institute in 1890.

FREDERICK FORRESTER HUNTING died at his home in Cardiff on April 27, 1935; he was sixty-two years of age. It was in 1923 that Mr. Hunting went to Cardiff from the Midlands to take over the management of the Cardiff Works of Messrs. Guest, Keen and Nettlefolds, Ltd., and he continued to hold that position until the amalgamation of the heavy-steel interests in South Wales of Messrs. Guest, Keen and Nettlefolds, Ltd., and Messrs. Baldwins, Ltd., when, with the closing-down of the Cardiff steelworks, he was appointed chief metallurgist to the British (Guest, Keen, Baldwins) Iron and Steel Co., Ltd. He joined the Iron and Steel Institute in 1917, to which, in collaboration with Mr. W. J. Brooke, he presented a note on "The microstructure of commercially pure iron between Ar_3 and Ar_2." Mr. Hunting was, at the time of his death, a member of the Corrosion Committee of the Iron and Steel Institute and the British Iron and Steel Federation.

Sir ROBERT MURRAY HYSLOP, a well-known industrialist, chairman of Messrs. Mosers, Ltd., iron and tool merchants, and a director of Messrs. Bayliss, Jones and Bayliss, Ltd., ironmasters, died on September 11, 1935, after a brief illness. He was educated privately and at University College, London, and joined Lloyd's as an underwriter. Subsequently he became associated with numerous industrial undertakings, and was soon a leading figure in the iron trade of the Midlands. For many years he lived at Beckenham, where he was prominent on the Urban District Council, on the Education Committee and in municipal affairs generally. He became a Justice of the Peace in 1909, and in 1917 he was knighted. A staunch Liberal, he was prominently identified with the Congregational Union and other Free Church movements, and with temperance work. Sir Murray was a Commander of the Legion of Honour, a Fellow of the Zoological Society, and since 1933 he had been a member of the Iron and Steel Institute.

JOHN KITCHING, a very old member of the Institute, died at Darlington on March 28, 1935, at the advanced age of eighty-one. For the greater part of his career he had been associated with the working of railways in Great Britain and in Spain. He joined the Iron and Steel Institute in 1881, and at his death was a life member.

WILLIAM WADDELL MCCOSH died at his home, Easter Moffat, Plains, near Airdrie, Lanarkshire, on March 17, 1935. Mr. McCosh was a son of the late Provost McCosh of Coatbridge and was educated at Cambridge. Later, he became one of the principals in the firm of William Baird & Co., Ltd., Gartsherrie Ironworks, and devoted most of his time to the interests of this concern. He was also a director of Baird's Mining Co., Ltd., and the Caledonian Portland Cement Co., Ltd. He served a period on the Coatbridge Town Council and was a magistrate when he retired from that body ; at the time of his death he was President of the North Lanark Unionist Association. He joined the Iron and Steel Institute in 1909.

THOMAS BROWN MACKENZIE, who died on August 5, 1935, was one of the leading engineers in the iron and steel industry. A native of Renfrew, he received his education in Glasgow and served an apprenticeship with Messrs. John Norman and Co., of Port Dundas. He obtained further experience at the Hyde Park Locomotive Works, Springburn, and returned to his former firm to take charge of the drawing office after the business had been taken over by Messrs. J. Copeland and Co. In 1895 he was invited to take charge of the Dalzell Works drawing office of

Messrs. David Colville and Sons, Ltd., and five years later he became chief engineer for all new construction. In this capacity he was responsible for the design and remodelling of plant at the various works of the company—such as the extensions at the Glengarnock and Clydebridge Steel Works, for instance.

He travelled widely in America and on the Continent in his study of steelworks practice, and was regarded as one of our ablest steelworks engineers. He was very much interested in the historical aspect of engineering, and was also deeply concerned with the welfare and education of young engineers; in furtherance of this object he founded a Mackenzie Trust Fund for the provision of awards to those who distinguished themselves at evening classes.

A life member of the Institution of Mechanical Engineers, a member of the Institution of Engineers and Shipbuilders in Scotland, and of the West of Scotland Iron and Steel Institute, he was also a life member of the Iron and Steel Institute, which he joined in 1901. Of his published papers one, on "Utilisation of waste heat from open-hearth furnaces for the generation of steam," was read before the Institute in 1918.

JOHN ALEXANDER MATHEWS, Vice-President of the Crucible Steel Co. of America, and a metallurgist of international repute, died of a heart attack at his home in Scarsdale, New York, U.S.A., on January 11, 1935. He was born at Washington, Pa., on May 20, 1872, and attended Washington and Jefferson College. In 1898 he took his Ph.D. degree at Columbia University and was appointed assistant in assaying and tutor in chemistry there. He was given the Barnard Fellowship and went to the Royal School of Mines, London, where he studied metallography under Sir William Roberts-Austen. Returning to Columbia in 1901, he worked on the development of alloy steels in Professor Howe's laboratory; in this he was encouraged by being awarded one of the first Carnegie Scholarships of the Iron and Steel Institute, in the year of their inception. The following year the first Carnegie Gold Medal was bestowed upon him; he thus achieved the signal distinction of being the first to receive either award and the first to be the recipient of the double honour.

He began his association with the Crucible Steel Co. in 1902, becoming metallurgist in charge of experimental work at Sanderson Brothers Steel Co., Syracuse. During this period, and subsequently, he became interested in steels for permanent magnets, and was instrumental in making the United States almost self-sufficient so far as tungsten and chromium magnet steels were concerned. He was of the firm opinion that the magnetic properties of carbon and alloy steels could ultimately be correlated with the structure and mechanical properties, and on this point

was responsible for many brilliant hypotheses—hypotheses which have since been amply confirmed

Dr. Mathews went to Halcomb Steel Co. in 1908 as operating manager, becoming President of this company in 1915. Five years later he was elected Vice-President of the Crucible Steel Co. of America, soon became President and only relinquished this post (in 1923) in order to become Vice-President and Director of Research, and so to devote more time to the latter. This position he held at his death.

Among his outstanding achievements were the use of the electric furnace in the production of high-quality steel, the American production of magnet steels, the development of various alloy steels, particularly high-speed steels containing vanadium, and the improvement of heat-resisting austenitic steels; in many of these fields he was the pioneer. During the war of 1914–1918 many technical committees benefited by his experience, especially in aircraft production problems.

Dr. Mathews was a member of many technical societies, including the American Society for Testing Materials and the American Society for Metals; he had been a member of the Iron and Steel Institute since 1904. A clear, lucid writer, he published about a hundred papers in all; those read before the Iron and Steel Institute include one on "A comparative study of some low-carbon steel alloys," in 1902, and one on "Retained austenite," in 1925. The reputation which he early established in Great Britain and America was further enhanced by the fact that he was chosen to deliver the Howe Memorial Lecture before the American Institute of Mining and Metallurgical Engineers in 1924, while in 1928 he was awarded the Robert W. Hunt Medal of that society.

Percy José Mitchell died in London on September 13, 1935, at the age of fifty-six. Born in London on May 15, 1879, he was educated at Stoke Newington. On leaving school he was articled to Mr. F. M. Rogers, but about a year later he was apprenticed to Mr. James Bennett, of Messrs. Bennett, Sons and Shears, Ltd., makers of sugar machinery and brewery and general engineers. In 1901, he was appointed London representative of Messrs. Fullerton, Hodgart and Barclay, Ltd., Paisley; while with them he installed plant in several tinplate works in South Wales and assisted in the erection of dock hoists and cranes, and refrigerating equipment. Two years later he started to practice as a consulting engineer. He acquired the patent rights of the Rateau system of exhaust steam utilisation in Great Britain, and first installed plant of this type for the Steel Company of Scotland; this was followed by the erection of similar plant in some fifty iron and steel works power stations, the exhaust

turbines in some cases being combined with steam accumulators. During the war Mr. Mitchell gave up his consulting practice and took up the manufacture of projectiles, aero engines and chemicals, the Staines Projectile Co. being formed for the purpose; nearly all the tools in the shell factory were constructed to his design. The war over, he turned his attention to the glass industry; he was concerned with the formation of the British Window Glass Co. to develop a Belgian patent, and also of Colonial and Foreign Glass Industries to handle the foreign patents covering this process. Further, he was responsible for the construction of a heavy cardboard and fibre-board mill at Merton for the Merton Board Mills, Ltd.; its erection presented difficulties owing to the marshy nature of the ground. In addition, he was concerned in the control of P. and B. Evaporators, Ltd., which manufactured evaporators constructed under the patents owned by Messrs. Prache and Bouillon, of Paris. Mr. Mitchell was a member of the Institution of Mechanical Engineers and the Institute of Sanitary Engineers; he joined the Iron and Steel Institute in 1905.

JOHN NEILSON, of Coatbridge, Lanarkshire, died early in July, 1935. At his death he was chairman of the Summerlee Iron Co., Ltd., of Glasgow. The Neilson family has been closely connected with the Scottish Iron Industry for a great many years, and it was J. B. Neilson, the uncle of Mr. John Neilson, who introduced the use of hot blast in the blast-furnace, which was probably the most marked advance in fuel economy made in the history of iron-making. Mr. Neilson was ninety-six years of age, and thus was almost certainly the oldest member of the Iron and Steel Institute. He was in fact the senior from the point of view of length of membership, for he was one of that small and dwindling group, the Original Members, which, with Mr. Neilson's decease, has now completely disappeared; thus, the last link with the days of the Institute's birth in 1869 has been broken.

STANLEY R. PATTERSON died at his home in Porthcawl on February 23, 1935, at the age of fifty-one. He started his professional career in May, 1897, when he joined the staff of Messrs. Guest, Keen and Nettlefolds, Ltd., as assistant to his father, the then chief engineer, whom he succeeded in 1918. When, about 1930, it was decided to close down the steel side of the works in Cardiff, he was moved to Port Talbot as chief engineer of the combined works of the British (Guest, Keen, Baldwins) Iron and Steel Co., Ltd., at Port Talbot and Cardiff, and he took in hand the important reconstruction work which has taken place at Port Talbot, culminating in one of the most modern installations

in Great Britain for the storage and use of blast-furnace and coke-oven gases in furnaces of all descriptions. He was also engaged on the preparation of the layout of an entirely new works to be built at Cardiff, and in the early part of 1934 it was decided to proceed with the new plant, a decision which necessitated the dismantling of the whole of the old plant at Cardiff. It is unfortunate that Mr. Patterson did not live to see the starting-up of the new works, the building of which had been his life's ambition and in the designing of which he had played so important a part. He had joined the Iron and Steel Institute in 1919.

WILLIAM HENRY WARDEN PROCTOR died at his residence in Warwick Road, Coventry, on December 18, 1934; he was fifty-eight years old. He was Managing Director of the Wycliffe Foundry Co., Ltd., Lutterworth, Rugby, of Follsain Metals, Ltd., and of the Globe Foundry, Coventry. He joined the Iron and Steel Institute in 1906.

CHARLES HENRY RIDSDALE, of Stoneycroft, Glaisdale, passed away on December 29, 1934, at the age of seventy-three. He first went to Tees-side in 1876, and there he became one of the first pupils in the laboratory of the late Dr. Stead. In 1879 he was associated with some of the early experiments carried out by Thomas and Gilchrist at the works of Messrs. Bolckow, Vaughan & Co., at South Bank, the outcome of which was the successful development of the basic Bessemer process, which made possible the manufacture of steel from local phosphoric ores. Later, Mr. Ridsdale was appointed chief chemist and technical adviser to the North-Eastern Steel Works, Ltd., of Middlesbrough. In 1917, he joined Messrs. Ridsdale & Co., analytical chemists, of Middlesbrough, and later became a director of and metallurgical consultant to the firm, in which his son, Mr. N. D. Ridsdale, now takes a principal part. He was a member of a number of technical societies, including the Chemical Society, the Institute of Chemistry and the Cleveland Institution of Engineers; he had joined the Iron and Steel Institute in 1883. He was the author of very many papers on technical matters, a number of which will be found in the *Journal of the Iron and Steel Institute*; in 1887, jointly with Dr. Stead, he presented a paper on "Basic slag, its formation, constitution and application, with special reference to crystals found therein," and then followed contributions on "Brittleness in soft steel," in 1895, on "Practical microscopic analysis for use in the steel industries, with an introduction to a systematic study of soft and dead soft steel," in 1899, on "The correct treatment of steel," in 1901, on "Diseases of steel," in 1903, and finally on "The valuation of ores and iron-making material," in 1920. In addition, he had collaborated

with his son, Mr. N. D. Ridsdale, in two papers presented to the Institute, one on "Mechanicalising analysis as an aid to accuracy and speed for commercial purposes," in 1911, and the other on "A new method for the accurate determination of phosphorus," in 1913. Mr. Ridsdale was the originator of standard analytical samples, and he had devoted much time and energy to their improvement and development.

RONALD JAMES HOWSON RITCHIE, of South Croft, Marton, Yorkshire, died in January, 1935, at the comparatively early age of forty. He was a director and works manager of the firm of James and Ronald Ritchie, cast-iron pipe manufacturers, and was indeed the son of the late Mr. James Ritchie who had founded the company. He was a member of the Council of the Middlesbrough Branch of the Institute of British Foundrymen. He took up membership of the Iron and Steel Institute in 1917.

GEORGE H. SANKEY, J.P., died at his residence, Astley Abbots, Bridgnorth, on December 19, 1934, at the age of sixty-seven. He was the second son of the late Mr. Joseph Sankey, the founder of the firm of Joseph Sankey and Sons, Ltd., of Bilston and Hadley (Shropshire)—now a branch of Guest, Keen and Nettlefolds, Ltd.—of which he became the managing director; he was connected with the Albert Street Works, Bilston, for about fifty years. He was High Sheriff of Staffordshire, and had served on the committee for the Wolverhampton and South Staffordshire district of the Royal Metal Trades Pension and Benevolent Society. He had been a member of the Iron and Steel Institute since 1903.

SECTION II.

NOTES ON THE PROGRESS OF THE HOME AND FOREIGN IRON AND STEEL INDUSTRIES.

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The Editor has been assisted in the preparation of these Notes by R. Elsdon, Librarian of the Institute.

MINERAL RESOURCES.

BRITISH EMPIRE.

The Canadian Mineral Industry in 1934. (Canada, Department of Mines, Mines Branch, 1935, Report No. 760).

Geology of the Grand Lake Coalfield. W. J. Wright. (Canadian Institute of Mining and Metallurgy, 1935, June, Bulletin No. 278, pp. 209–216). The coalfield is situated in Queens and Sunbury Counties, New Brunswick, Canada, chiefly within a three-mile radius of Minto. The seam varies in thickness from 16 to 30 in. and lies almost horizontal, at depths of 0 to 125 ft. The coal is sold in four grades: Screened or lump, run of mine, slack, and stripping. The chief market is with the railways and commercial plants, with a fair amount for domestic purposes. The coal is bituminous, possessing good coking qualities, but rather high ash and sulphur contents. The following is the proximate analysis: Moisture at 105° C., 1·1; volatile matter, 31·4; fixed carbon, 56·0; ash, 17·0; sulphur, 7·1 per cent. The calorific value is 12,840 B.th.u. The author describes the geological sequence of the coal-bearing strata, and the more interesting features connected with the rock formations.

A Study of the Barakar Coals of the Jharia Coalfield, Bihar and Orissa, India. C. Forrester. (Journal of the Institute of Fuel, 1935, vol. 9, Oct., pp. 30–58). An account is given of the geology of the district, the chemical and physical characteristics of the Jharia coals, and the nature and distribution of the mineral matter. Washability curves of the coal are also included.

The Natural Gas Resources of Burma. C. T. Barber. (Memoirs of the Geological Survey of India, 1935, vol. 66, Part 1).

EUROPE.

Gold, Tin and Tungsten in the Montagne de Blond (Haute-Vienne). (Génie Civil, 1935, vol. 107, Sept. 28, pp. 301–302). A short note on the geology of this district in France.'

The German Iron Ore Deposits and their Possible Utilisation. G. Einecke. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Sept. 14, pp. 1099–1110). The author presents a review of the iron ore deposits of Germany and considers their usefulness and the question of how long they will last out. He

discusses the total reserves of the country and the possibility of increasing the output in the individual districts; from this he draws conclusions as to whether such increased outputs will suffice for Germany's need of ores and what consequences would follow on an excessive increase in the output.

Iron Ore Reserves of Germany. E. Kohl. (Archiv für die Lagerstättenforschung, 1934, No. 58: Glückauf, 1935, vol. 71, June 22, pp. 590–592). The results of a survey of the iron ore reserves in all the ore deposits in Germany, carried out in 1931 and 1932, are presented.

The Formation of the Siegerland Spathic Iron Ore Deposits by Lateral Secretion. H. Breddin. (Glückauf, 1935, vol. 71, Aug. 31, pp. 821–830).

The Oolitic Iron Ores in the Dogger Sandstone of the Franconian Jura. E. Schmidtill. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, July, pp. 1–13). After describing the petrographic and chemical composition of these iron ores, the author outlines the working conditions in the various districts and discusses the preparation and smelting of the ores.

The Iron Ore Deposits of the Eisenwerkgesellschaft Maximilianhütte, Rosenberg. N. Hamacher. (Stahl und Eisen, 1935, vol. 55, June 27, pp. 693–695). The development, extent and mode of occurrence of these iron ore deposits are described and the question of the preparation of the ore is discussed. The estimated reserves and proposed annual outputs suggest that the mines opened up so far will have a limited life. The ore field lies to the east and north of Nuremberg.

The Ores and Metallurgical Industries of Rumania and Czechoslovakia. C. Berthelot. (Revue de Métallurgie, Mémoires, 1935, vol. 32, Apr., pp. 156–170; May, pp. 200–209).

The Mining and Metallurgy of Sardinia. C. Berthelot. (Revue de Métallurgie, Mémoires, 1935, vol. 32, Aug., pp. 342–350). A brief review of the principal mining and metallurgical activities in Sardinia.

A Chromite Deposit at Tschatschak, South Slavia. G. Petunnikov. (Montanistische Rundschau, 1935, vol. 27, Aug. 1, pp. 1–3). Brief details of the deposit are given.

AMERICA.

Geology of Big Horn County and the Crow Indian Reservation, Montana, with Special Reference to the Water, Coal, Oil and Gas Resources. W. T. Thom, jun., G. M. Hall, C. H. Wegemann and G. F. Moulton. (United States Geological Survey, 1935, Bulletin No. 856).

ASIA.

The Mineral Industry of Japan. (Colliery Guardian, 1935, vol. 151, Aug. 23, pp. 329–331). This review includes data on the production of minerals and the character of the coal mined in Japan.

Geology of the Iron Deposits in the Lower Yangtze Region. C. Y. Hsieh. (Geological Memoirs, National Geological Survey of China, 1935, Series A, No. 13).

ORES—MINING AND TREATMENT.

Concentration of Titaniferous Iron Ore from Taberg, Sweden, after Previous Heating. G. G. Bring and P. G. Kihlstedt. (Jernkontorets Annaler, 1935, vol. 119, No. 8, pp. 303–342). Ore from Taberg is of the titaniferous type in which the magnetite grains show lamellar intergrowths of ilmenite ; the lamellæ are so thin that concentration by mechanical means is not possible. Furthermore, a rim of tough hornblende binds the magnetite to the gangue (chiefly olivine), making separation difficult. Attempts were made to effect a separation by heating the ore to temperatures up to 1,100° C. and quenching it in water. Heating to 1,000° C. renders the material friable, but at the same time magnetite is formed in quite large quantities in the hornblende, making the gangue strongly magnetic. On the other hand, the magnetite grains themselves become more feebly magnetic. Magnetic separation then becomes difficult, and heating does not, therefore, form a solution to the problem.

Roasting and Sintering of Ores in a Turbulent Furnace. M. Fourment. (Revue de Métallurgie, Mémoires, 1935, vol. 32, June, pp. 245–247). After touching on various systems of roasting or sintering ores, the author describes the Saint-Jacques furnace. In this the powdered material is injected under pressure in a circular manner into the top of a cylindrical chamber lined with refractory. At the bottom are a number of burners arranged tangentially ; the flames develop in a spiral manner and the waste gases are evacuated through a central opening at the top. Above the burners are placed tuyeres, through which hot air (to prolong the flames or to cause oxidation) is blown. The particles of ore thus traverse a form of spiral from the top to the bottom in a turbulent atmosphere. As the waste gases are hot and carry valuable dust, they pass through a recuperator and a dust extractor (cyclone) ; the recuperator supplies the hot air for the tuyeres. Such a plant has been erected in Eastern France for the sintering of blast-furnace dust and pulverulent iron ores ; this furnace is fired with coke-oven gas, and a temperature of 1,400° C. is reached. The material collects in a hopper at the bottom, and sinters together by simple contact ; the sinter produced, without pressure, is porous and hard, and is very suitable for charging into the blast-furnace.

Progress in the Sintering and Sinter-Roasting of Ores. H. Wittenberg. (Periodic Review of the Metallgesellschaft, 1935,

May, No. 10, pp. 11–16). The merits of the various types of apparatus operating on the Dwight-Lloyd system, namely, stationary sinter pans, circular rotating pans and straight-line machines, are compared. Some examples of the application of these sintering machines in the roasting of iron ore and sulphide ores are cited.

The Ore Loading Plant at the Harbour of the Ilseder Hütte on the Mittelland Canal. (Demag News, 1935, vol. 9, July, pp.30–31). The main features of the ore loading equipment at the harbour of the Ilseder Hütte are a mobile loading belt and a number of self-discharging ore cars. The cars discharge the ore on to the belt, which is carried by a travelling gantry, and the belt delivers the ore into adjacent barges. In this way a cargo of 2,500 to 3,000 tons of ore daily is dealt with.

REFRACTORY MATERIALS.

Utilisation of French Refractories in Metallurgy. D. Petit. (Revue de Métallurgie, Mémoires, 1935, vol. 32, July, pp. 282–292). After touching generally on the properties of refractories and methods of testing them, the author goes into more detail regarding the principal refractory materials used in metallurgy (siliceous, silico-aluminous, aluminous materials, so-called "super-compressed" products, special refractories, magnesia-base refractories, insulating bricks, lining material, cements and mastic), and concludes by giving a few examples of their use. His purpose is to show the importance of the French refractory industry.

Difficulties in the Manufacture of Firebrick. C. R. F. Threlfall. (Sands, Clays, and Minerals, 1935, vol. 2, June, pp. 67–70). The difficulties which are met with in the manufacture of firebricks are chiefly these: Lack of uniformity in the raw materials; variation in humidity; unequal mixing of the ground clay and water. Vacuum mixing has been introduced in order to overcome the latter trouble. Drying no longer presents as many difficulties as was once the case, but further progress is both possible and desirable.

Firing Clay Refractories with By-Product Coke-Oven Gas. C. E. Bales and W. T. Christian. (Bulletin of the American Ceramic Society, 1935, vol. 14, Aug., pp. 245–250). The kiln and pipe-line equipment and firing procedure adopted for firing clay refractories with by-product coke-oven gas are described. The gas has a calorific value of 550 B.th.u.; it is stated to be more economical than natural gas, and there is no difficulty in reaching the required high temperatures. A better-looking and more uniform product is secured by the use of this fuel than by the use of coal.

A New Application of the Ritex Process to the Manufacture of Chrome Brick. H. P. Heuer. (Steel, 1935, vol. 97, Aug. 19, pp. 22–25, 30). Ritex chrome brick is compared with ordinary chrome brick in respect of chemical analyses, density, refractoriness, crushing strength and volume stability. It is shown that Ritex chrome brick is superior to ordinary burned chrome brick, besides being more resistant to spalling and having a lower heat conductivity. Ritex contains about 32 per cent. of magnesite, compared with about half this amount in ordinary chrome brick; tests revealed that the magnesite present was to some extent responsible for the improved physical characteristics. During manufacture

Ritex brick undergoes sizing and bonding processes, but no kiln firing is adopted. As a substitute for burned chrome brick, or, in certain situations, for silica and magnesite brick, Ritex is claimed to be entirely successful.

Magnesite and Magnesite Bricks. C. L. Rigby. (Sands, Clays, and Minerals, 1935, vol. 2, June, pp. 99–104). Supplies of magnesite are largely derived from Austria, Greece, Canada, British India, and Soviet Russia. The mineral is of two kinds, amorphous or crystalline. The former is used for building purposes, since it does not readily sinter; the crystalline variety contains impurities which induce vitrification when burnt at sufficiently high temperatures, and this renders it suitable for refractories. The raw magnesite is dead-burned at 1,550°–1,600° C. in rotary kilns. It is allowed to disintegrate in storage bunkers, and is afterwards treated in mechanical separators, followed by grinding and screening. It is dampened with water and allowed to hydrate, then pressed into bricks, which are dried and burnt.

Forsterite and Other Magnesium Silicates as Refractories. R. E. Birch and F. A. Harvey. (Journal of the American Ceramic Society, 1935, vol. 18, June, pp. 176–192). In a survey of magnesium silicates for use as refractories, experimental studies on talc, serpentine, olivine, and mixtures of these minerals with magnesia, are reported. Forsterite ($2MgO.SiO_2$) is the only stable magnesium silicate capable of withstanding elevated temperatures. Olivine always contains talc and serpentine, and these provide a low-melting bond which causes the refractory to fail readily at high temperatures. If MgO is incorporated, however, the accessory minerals are converted to forsterite, and the resulting refractory possesses satisfactory properties. These properties—including refractoriness, volume stability, thermal expansion, chemical behaviour, &c.—are discussed.

Dust Eliminated in Crushing Department of Inland Steel Co. (Iron Age, 1935, vol. 136, Aug. 22, pp. 28–29, 67). In grinding the refractory material at plant No. 1 of the Inland Steel Co., Indiana Harbour, Ind., U.S.A., a considerable amount of dust is produced. The methods of dealing with this dust are described.

The X-Ray Detection of Tridymite in Silica Coke-Oven Bricks and its Technological Importance. K. Endell, U. Hofmann and E. Maegdefrau. (Glückauf, 1935, vol. 71, Sept. 7, pp. 862–863). Koeppel has indicated that tridymite brick (that is, silica brick containing as much as possible of tridymite crystals) is the most suitable material for the construction of coke-ovens. The present authors touch on the detection of tridymite by means of X-rays, tabulate the properties of a so-called 100 per cent. tridymite brick and of a silica coke-oven brick made in the normal way, and indicate the inferences to be drawn.

An Apparatus for Investigating the Corrosive Action of Slags upon Refractory Materials. A. E. Vickers and R. A. Bell. (Journal of the Society of Glass Technology, 1935, vol. 19, June, pp. 150T–155T). An apparatus for determining the resistance of refractory materials to the action of slags at high temperatures is described. It consists of a circular furnace lined with the refractory materials under test, and heated by a rotating blowpipe supplied with gas and air under pressure. The slag is fed through the blowpipe flame, and impinges on the heated refractories. The slag, in powdered form, is supplied continuously or intermittently. Some results on various refractories, using powdered glass as the slag, are included.

Refractories for Foundry Use. W. J. Rees. (Institute of British Foundrymen, July 2 to 5, 1935 : Foundry Trade Journal, 1935, vol. 53, July 18, pp. 47–48). The considerations which determine the type of refractory used in the cupola, the conditions to which it is subject, and the properties which it is desirable that the materials should possess, all receive attention. The behaviour of rammed linings and of patching materials is briefly dealt with.

Notes on the Texture of Refractories. Part I.—Non-Regularity of Texture of Gasworks Fireclay Refractories and Its Possible Effect on Durability. T. F. E. Rhead and R. E. Jefferson. **Part II.—Pictorial Methods of Recording the Texture of Refractories or Similar Materials.** T. F. E. Rhead, J. N. Shorrock, and C. L. Evans. (Transactions of the Ceramic Society, 1935, vol. 34, Aug., pp. 363–379). In Part I. the authors make notes on cases of irregular texture which have come to their notice. Laminations they regard as injurious, being likely to give rise to slag penetration, contraction and spalling. Improper pugging or forming of the clay mixture is shown to be responsible for many cases of poor texture. In Part II. methods of preparing the surface of refractory brick to enhance the contrast between solids and voids and obtain a permanent record are described. The most successful appears to be that of forcing in black molten wax under 30 atm. pressure of nitrogen. In another method ink impressions are suggested as a substitute for photography. The brick is ground, inked with printer's ink, faced with roller composition, and an impression is taken on a similar slab of roller (gelatin) composition. From this slab a print is taken. The remarks of the authors of each section are illustrated by photographs.

Coke-Oven Door Brick. C. Koeppel. (Glückauf, 1935, vol. 71, Aug. 24, pp. 797–805 ; Aug. 31, pp. 830–834). The causes of the destruction of coke-oven door brick lie in sharp temperature changes and the precipitation of carbon within the material as a result of the decomposition of carbon monoxide. As the brick

on ovens with ends which are kept hot seldom cools below the critical temperature at which the expansion of silica suddenly alters, the ability of the brick to withstand rapid temperature variations no longer has the importance in modern ovens that it had in older fireclay ovens. Of more importance to the life is the ability of the brick to stand up to the deposition of solid carbon. The destruction of a fireclay brick, which proceeds in separate stages, can be well followed in door bricks, which permit each stage to be observed. There are several ways of increasing artificially the resistance of refractory bricks to the action of solid carbon: (1) Their decomposing effect on carbon monoxide may be reduced; (2) their structural resistance to the mechanical force of the carbon may be increased; (3) the temperature zone within which iron oxides accelerate the decomposition may be shifted to deeper-seated and less-endangered layers in the brick. The means necessary to these ends often influence adversely the resistance to sudden cooling, but, nevertheless, there are ways of maintaining this property at sufficiently high values. As a result of the close connection between the maximum carbon monoxide decomposition and the temperature interval between 400° and 600° C., in which iron oxides assist the breakdown of the gas, door bricks which are kept hot last longer than those on cold doors or oven ends. Thus there is a relationship between the temperature of the end flue and the life of the brick lining of the door. Ovens which are worked cool set the most difficult conditions, and these can be adequately met only by the very best door brick material. The production of such door bricks is difficult, and requires that consideration must be given to all the peculiarities of coke-oven operation.

As simple tests for the suitability of door bricks the following may be used: Thermal expansion test, together with the determination of the resistance to change of temperature and the proportion of mullite present; further, the determination of the oxide form of the iron contained in the fireclay. If a distinction be drawn between hot ovens (end flue temperature over 1,200° C.), medium-warm ovens (1,100°–1,200°) and cool ovens (end flue temperature below 1,100° C.), it is possible to draw up graded qualities which will satisfy the three conditions.

Malleable Furnace Refractories. L. C. Hewitt. (American Foundrymen's Association, Aug., 1935, Preprint No. 13). A review of recent developments in malleable iron furnace refractories and their applications.

Refractory Insulating Bricks as Structural Materials for Modern Reheating Furnaces. E. Senfter. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, May, pp. 473–478). In non-continuous reheating furnaces—that is, those which are heated up during the heating of each charge and then cool off more or less before the next one—the total heat requirement is made up of that

needed for heating the material plus the wall losses and the heat absorption losses. In fuel-fired furnaces, the influence of the heat absorption is not indicated by the heat balance but by the distribution of the heat requirements of the furnace. Investigations on reheating furnaces have shown that the heat absorption and wall losses account in general for 60 to 80 per cent. of the total fuel consumption. The extent of heat absorption is determined by the weight of the structural material; insulating bricks are light-weight refractories, and they offer the combined advantages of low heat absorption and low wall losses. Their use in part or in whole for furnace structures brings considerable economies. For ordinary reheating furnaces, in which the heating period is about 20 hr. and the temperature 700° to 900° C., a change-over from the usual construction to the light form with insulating bricks may effect fuel savings of 50 per cent. or more. The cost of these light insulating refractories per cubic metre of wall is not much greater than that of ordinary construction.

Modern Practice in Heat Insulation. C. Ellis. (Engineering and Boiler House Review, 1935, vol. 49, Sept., pp. 170–176). A review of the chief insulating materials available with some notes on their respective applications.

Insulating-Refractory Lined Furnaces. J. G. Coutant. (Industrial Heating, 1935, vol. 2, Sept., pp. 522–524). It is shown that the replacement of fire-clay by insulating-refractory linings is likely to reduce the radiation losses and thermal capacity while increasing the efficiency of the furnace.

Thermal Insulation of O.-H. and Electric Furnaces. (Foundry Trade Journal, 1935, vol. 53, July 25, pp. 60, 74). This article is an English translation of an article on the above subject by S. Wohlfahrt and O. Ell published in Jernkontorets Annaler, 1934. (*See* Journ. I. and S.I., 1935, No. I. p. 327.)

Insulation and Control of Open-Hearth Furnaces. W. C. Buell, jun. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 641). This paper is in two parts. The first deals with the insulation of open-hearth furnaces. It is pointed out that if the full benefit of insulation is to be reaped the furnace must be of suitable design. The author proceeds to classify insulating materials as follows: Class *A*, comprising semi-refractories able to withstand loads at high temperatures; Class *B*, comparatively weak insulating bricks for moderate temperatures; Class *C*, comprising loose materials such as slag wool; Class *D*, which includes cements and plasters; and Class *E*, which includes concrete. He then briefly examines heat problems in which insulating materials are involved. The second part of this paper deals with the control of incoming fuel and of outgoing furnace gases. Methods of control (manual or mechanised) are also dealt with. Finally, the author makes one or

two practical suggestions on the improvement of open-hearth furnaces.

Glass Silk for Insulation of Heat and Sound. A. L. Forster. (Transactions of the Liverpool Engineering Society, 1935, vol. 56, pp. 159–174). This paper includes an account of the nature of glass silk and of its properties as a heat insulator compared with those of other heat insulating media.

FUEL.

PROPERTIES, PREPARATION, MANUFACTURE AND USES.

FUEL TECHNOLOGY.

Temperature Correction in the Calorimetric Determination of the Calorific Value. W. Schultes and R. Nübel. (Wärme, 1935, vol. 58, p. 15: Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Oct. 5, pp. 1207–1208). The authors discuss old formulæ and suggest new ones for correcting the temperature when using a calorimeter, to allow for the influence of the exchange of heat between the calorimeter and its surroundings.

Twin Calorimeter for Small Heat Evolutions. H. Gawlick. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Sept. 7, p. 1089). Brief illustrated particulars of this calorimeter are given.

A Furnace for the Measurement of Coal-Ash Softening Temperature. E. P. Barrett. (American Society for Testing Materials, 1935, Preprint No. 74). The author describes a furnace designed for the specific purpose of determining coal-ash softening temperatures and which is readily adaptable to measuring cone deformation temperatures up to 1,650° C. The furnace is of the horizontal muffle type and is so constructed as to permit the cones to be viewed against a background at a somewhat lower temperature, so that they are readily visible at all temperatures between 815° and 1,650° C. The advantages of this type of furnace over the conventional pot furnace are emphasised.

Control of Forms of Iron in the Determination of Fusion Temperatures of Coal Ash. W. T. Reid. (Industrial and Engineering Chemistry, 1935, Analytical Edition, vol. 7, Sept. 15, pp. 335–338). The degree of oxidation of the iron in coal-ash slags affects the temperatures at which the various stages of fusion occur; hence control of oxidation is necessary in investigating the properties of such slags. The author devised an electric furnace which enabled the ash sample to be melted in an atmosphere of nitrogen; in this way oxidation of the iron was avoided. Determinations of the relation between flow temperature and the degree of reduction of the iron were made, the results of which are given.

Observations on the Fusibility of Coal and Coke Ash. S. Tweedy. (Coke Oven Managers' Association, Sept. 10, 1935: Gas World, 1935, vol. 103, Oct. 5, Coking Section, pp. 9–14).

Principles, Development and Examples of Thermotechnological Calculations. Part III.—Basis and Development of Formulæ for the Calculation of the Heat given up in the Furnace by the Fuel Gases. H. Schwiedessen. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, July, pp. 23–30). The present paper is a sequel to two earlier ones (*see* Journ. I. and S.I., 1935, No. I. p. 328). The pre-calculation of changes in the fuel consumption that will arise out of changes in the construction or operation of a furnace is usually very difficult, and generally can only be done on the basis of the results of experience. Nevertheless, there are a few cases, where certain definite assumptions may be made, in which it can be done without the aid of such data; they arise when the heat imparted to the charge, the cooling water and the walls of the furnace is the only factor that is varied. Practically the only way to alter these heat quantities is by varying (*a*) the combustion arrangements, (*b*) the preheating, or (*c*) the calorific value of the fuel. The author deals with the calculation of the change of fuel consumption under these conditions and gives examples.

Combustion. H. A. Fells. (Yorkshire Junior Gas Association, Apr. 13, 1935: Gas World, 1935, vol. 102, May 18, Industrial Gas Supplement, pp. 10–14). The author enumerates the principal factors which govern the application of town gas to industrial heating, more especially with regard to the process of combustion. The optimum air/gas ratio, the influence of preheating, and the production of luminous flames, are considered in turn. The process of combustion is then dealt with, reference being made to the nature of the furnace atmosphere and of the combustion products.

Fundamental Combustion Data Trace Path to More Efficient Use of Industrial Gas. E. O. Mattocks. (Steel, 1935, vol. 96, June 10, pp. 30–33, 64). Some of the points which emerge from researches carried out by the American Gas Association are recounted. The author outlines the elementary theory of gas combustion, and describes the flames obtained by the use of fully pre-mixed air, partly pre-mixed air, and unmixed gas flames. Curves showing the amount of oxygen necessary for complete combustion are reproduced. The author points out that, if necessary, combustion chambers could be made only just large enough to house the flame; he therefore suggests that a combustion chamber, a heat exchanger and a work chamber should all be used. Finally, the types of burner employed are dealt with.

Influence of Dissociation on the Temperatures of Flames. L. Reingold. (Chaleur et Industrie, 1935, vol. 16, June, pp.

275–284; Oct., pp. 463–472). The author presents a mathematical study of the influence of dissociation on the temperature of flames. He has obtained systems of equations of which the nomographic solutions enable the determination of the decrease of temperature from the theoretical temperature of combustion due to the effect of dissociation and also of excess air.

Doctrine Relative to Flame, Its Characteristic Temperature and Its Radiation. G. Ribaud. (Chaleur et Industrie, 1935, vol. 16, June, pp. 291–294). The author discusses the following points: Definition of flame; choice and definition of the theoretical temperature of combustion of a flame; physical data for gases, usual constituents of flames; calculation of the theoretical temperature of combustion; and the part played by the radiation of gases in the thermal exchanges within industrial furnaces.

Principles of Fuel Beds. P. Nicholls. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 629: Colliery Guardian, 1935, vol. 141, Aug. 2, pp. 191–193; Aug. 9, pp. 237–238). The types of fuel bed may be classified as follows: (1) Underfeed, in which the fuel and air move in the same direction; (2) overfeed, in which the fuel and air move in opposite directions: (3) crossfeed, in which the fuel and air move at right angles to each other. Curves may be drawn which enable the rate of ignition and rate of burning to be compared with the thickness of the bed and rate of air supply. The distribution of ash and coke in the fuel beds differs with the method of feed; curves may be plotted to demonstrate the differences between the underfeed and the overfeed types.

Controlled Luminous Flame Affords Uniform Heating. J. B. Nealey. (Steel, 1935, vol. 97, Sept. 9, p. 39). A short note on the principle of controlled luminous flame heating is furnished. Gas and air are allowed to enter at ports in the furnace, the supply being regulated by valves; control over the velocity of the gas stream is established by introducing a mixture of air and waste gas. In this way the velocity of the furnace gases and the length of flame may be regulated.

The Development of Convection Heating Equipment. (Industrial Heating, 1935, vol. 2, Sept., pp. 476–492). A short discussion of the principle of convection heating is followed by a review of the chief items of furnace equipment employed. Attention is devoted to the types of furnace (batch or continuous), the method of heating (direct firing, indirect firing, and electric), the fans and exhausters, and the conveyors, &c., for handling the materials.

Practice in the Construction of High-Pressure Boilers. W. Seeberger. (Stahl und Eisen, 1935, vol. 55, Aug. 15, pp. 877–882). The author outlines the development of high-

pressure boiler plants for pressures of 60 and 120 atm., and describes experiences gained. The influence of the capacity, pressure and steam temperature on the design and construction of modern high-pressure boilers is shown by means of examples.

The Development of the Benson System of Steam Raising. H. Gleichmann. (Stahl und Eisen, 1935, vol. 55, Aug. 29, pp. 930–932). The author indicates the best materials to use in the construction of various parts of the Benson boiler, and discusses the convertion of the power plant in metallurgical works to high-pressure operation.

Twenty-Five Years' Progress in the Scientific Control of Steam Boiler Plant. D. Brownlie. (Steam Engineer, 1935, vol. 4, June, pp. 368–382). The development of boiler control devices is sketched; CO_2 recorders, boiler-feed meters, steam meters, coal weighing machines, oil and gas meters, and pyrometers are included in the instruments described.

Waste-Heat Boilers in Open-Hearth Practice. (Iron and Steel Institute, Sept., 1935, Special Report No. 10). The report contains five sections. Section *A* is introductory, while Section *B* is a report of a waste-heat boiler trial at the works of the Park Gate Iron and Steel Co., Ltd. This includes a description of the furnace and of the methods of taking measurements, and an analysis of the results. The section which follows deals critically with the economics of waste-heat boiler operation, with particular reference to the influence of the furnace and regenerators. Section *D*, by A. F. Webber, deals with the design, installation and operation of waste-heat boilers; among the subjects considered are: Alternative water-tube and fire-tube designs; the use of economisers; the use of air heaters; infiltration; heat insulation; feed-water treatment; heat input; and boiler tube design. The final section is a historical summary of waste-heat boiler practice, to which is appended a bibliography.

Waste-Heat Boilers in Gas Works Practice. W. Gregson. (Institution of Gas Engineers, June, 1935: Fuel Economist, 1935, vol. 10, July, pp. 921–927: Iron and Coal Trades Review, 1935, vol. 130, June 14, p. 1013). The problem of recovering heat from flue gases from the retort settings is discussed; it is shown that there are limitations to the use of regenerators, and that if these are eliminated the heat may be effectively recovered in waste-heat boilers. The latter are dealt with in detail, particularly the respective merits of water-tube and fire-tube boilers and the efficient operation of the latter.

Scientific Management as an Aid to Fuel Efficiency in the Iron and Steel Industry. S. L. Poole and E. C. Evans. (Fuel Economy Review, 1935, vol. 14, pp. 18–19). The ways in which efficiency may be improved and wastage reduced are briefly considered.

Some Modern Temperature Control Apparatus. (Engineering and Boiler House Review, 1935, vol. 49, July, pp. 26–38). Some of the leading types of temperature control equipment are described and illustrated. These include control pyrometers, regulators of the thermo-relay type, and those depending on mercury expansion or thermostatic operation.

Open-Hearth Temperature Control. E. Cook. (American Society for Metals, Oct., 1935, Preprint No. 27). The calibration and use of optical pyrometers in steel making are discussed. A working system of temperature control for the open-hearth furnace is suggested with reference to the limitations of instruments and melting practices. Operating data are presented, together with such items of plant practice as may be benefited by successful application of the methods.

Automatic Control over Metallurgical Furnaces. J. K. Mawka. (Blast Furnace and Steel Plant, 1935, vol. 23, July, pp. 475–476). The author suggests that it is desirable to adopt automatic control of the furnace pressure and the fuel/air ratio.

Recent Advances and Trend of Developments in Gas Measurement in Europe, as compared with America. R. Leonhardt. (Gas World, 1935, vol. 103, Sept. 7, pp. 188–192, 195). Modern gas metering practice in Europe is compared with similar practice in the United States.

Organisation of the Gas-Volume Measurement at the Friedrich-Wilhelms-Hütte, Mulheim. W. Hähn. (Stahl und Eisen, 1935, vol. 55, Sept. 19, pp. 1013–1015). The system of record cards used for calculating and recording the readings of gas measurements used at this works is described.

The Trend of Electrification in the Steel Plants of the U.S.S.R. P. L. Litvin. (Iron and Steel Engineer, 1935, vol. 12, May, pp. 290–292). An outline of the progress of electrification in Soviet steel plants is given. Before the war steel plants were operated by manual and (occasionally) hydraulic or steam appliances. Since the war and the inception of the five-year plan, electric and automatic operation have been extensively adopted.

A Graphical Method of Calculating Heat Loss through Furnace Walls. C. E. Weinland. (Transactions of the American Society for Metals, 1935, vol. 23, June, pp. 431–454).

Heat Transmission in Some Types of Boilers. P. St. G. Kirke. (Journal of the West of Scotland Iron and Steel Institute, 1935, vol. 42, Feb., pp. 77–86). The author considers the rates of heat transmission in fire-tube and water-tube boilers. He then describes some results obtained with the use of " Sinuflo " fire-tubes. These tubes are sinuously shaped along the length of the tube. However, they are easy to clean, and the boiler efficiencies are high.

Study of Continuous Recuperators—Calculation of Heat Exchangers. Laffargue. (Chaleur et Industrie, 1935, vol. 16, July, pp. 339–344). A mathematical discussion of the characteristics of heat exchangers.

Note on Cross-Current Heat Exchangers. A. Liebaut. (Chaleur et Industrie, 1935, vol. 16, June, pp. 286–290). The author has developed formulæ by means of which the thermal characteristics of cross-current heat exchangers can be calculated.

COAL.

British Coals, their Analyses and Uses. (Fuel Economist, 1935, vol. 10, July, pp. 946–947 ; Aug., pp. 977–979). Details which include the proximate analysis, sulphur content, calorific power and evaporative power, of a number of British coals are set forth, together with the names under which they are marketed. (*See* Journ. I. and S.I., 1935, No. I. p. 334).

Coal and Capillarity. P. Nashan. (Glückauf, 1935, vol. 71, Aug. 24, pp. 805–812). After describing the principle of the method employed, the author describes tests made to determine the capillarity of coals from various seams. The results show that the capillarity index is probably characteristic for each seam, and that seams can be collected together according to their capillarity into groups which correspond to the subdivisions based on the age of the coals.

Studies in Coal by X-Ray Diffraction Methods. C. Mahadevan. (Quarterly Journal of the Geological, Mining and Metallurgical Society of India, 1935, vol. 7, No. 1 : Fuel in Science and Practice, 1935, vol. 14, Aug., pp. 231–241). A brief introduction to the subject of X-ray analysis is followed by a description of the experimental procedure followed in studying coal samples. The author then proceeds to give the results of his investigations. He shows that vitrain, fusain and durain are each responsible for different varieties of diffraction pattern. The scattering index was found to be correlated with the sum of the moisture content and volatile matter rather than either individually. Experiments on dehydrated and volatile-free (presumably coked) coals suggested that scattering may be attributed to the presence of moisture and volatiles. The alpha, beta and gamma compounds obtained by benzene extraction were studied, and the two former found to be nearly identical. In vitrain the ash appeared to be present in the colloidal form, and in durain it was partly colloidal and partly suspended ; in fusain the ash was in the coarse form. The results of comparative tests on coals of different geological periods are also reproduced.

Importance of Pulp Density, Particle Size and Feed Regulation in Flotation of Coal. J. T. Crawford. (American Institute of Mining and Metallurgical Engineers, 1935, Contribution No. 86).

The Shape and Specific Surface of Coal Particles. L. W. Needham and N. W. Hill. (Fuel in Science and Practice, 1935, vol. 14, Aug., pp. 222–230). The authors consider the problem of calculating the specific surface of small particles on the basis of the conception of the " statistical mean shape." For this purpose the parallelism of the surfaces was evaluated (by observation) and an examination of particle silhouettes was carried out. This enabled the surface area to be computed, and the values were compared with those obtained by a direct method—taking a mould in plasticine. The results, on particles from 1½ in. down to ½ in. in size, were found to be in good agreement, and it is suggested that the statistical method may be applied to determinations involving much smaller particle sizes. The authors discuss the reliability of the formula used, on the basis of an examination of different types of coal.

Innovations in Coal Preparation, 1933 and 1934. A. Götte. (Glückauf, 1935, vol. 71, June 15, pp. 553–562; June 22, pp. 582–590; June 29, pp. 601–609). A complete review of all aspects of innovations introduced into the preparation of coal during the years 1933 and 1934. Very numerous references to the literature of the subject are given.

Note on the Estimation of Fusain in Coal Dusts. R. Belcher. (Fuel in Science and Practice, 1935, vol. 14, June, pp. 181–182). The fine dust obtained from de-dusting coals may be usefully admixed with coal before coking. The improvement of the quality of the coke may be attributed to the presence of fusain. The amount of fusain in the dust may be determined by oxidation and subsequent filtration; for the latter purpose it is convenient to use an immersion filter, and this the author describes. The process of filtration is very much expedited by its use.

The Selection of a Coal-Cleaning Plant. J. N. Williamson. (Colliery Engineering, 1935, vol. 12, July, pp. 227–229). It is shown that the selection of a coal-cleaning process depends primarily on the nature of the washability curves of the coal; in this connection, the character of the middlings product is usually the factor which determines the process employed. The physical character of the principal impurities, the absence or presence of salt, and the amount of surface moisture on the mined product are among the factors which govern the choice of wet or dry systems. It would appear that a combination of wet and dry processes is likely in many cases to prove the most successful.

Bituminous Coal Cleaning in the United States. (Iron and Coal Trades Review, 1935, vol. 130, May 31, p. 946). This article

is based on a statistical survey of the methods adopted for cleaning American bituminous coals.

The Grading and Classification of Coal. H. E. Mitton and D. T. Davies. (Institution of Mining Engineers, July 11, 1935: Colliery Guardian, 1935, vol. 151, July 12, p. 63). It is the opinion of the authors that the market for large coal is declining, and that the breaking and grading of coal are likely to become increasingly important. They suggest further, that greater care must be taken in preventing breakage in the transport and handling of the coal.

The Pneumatic De-dusting of Coal. S. R. Berrisford and R. H. Allen. (North Staffordshire Institute of Mining Engineers: Colliery Guardian, 1935, vol. 151, July 12, pp. 55–58; July 19, pp. 99–100). The authors lay down certain conditions which they consider an ideal de-dusting plant should fulfil, and proceed to examine the salient characteristics of modern de-dusting plants. De-dusting efficiency is then treated; data derived from a de-duster test are furnished, and the efficiency curves are discussed. Finally, the effect of moisture and ash on the efficiency of operation are considered, together with the results of de-dusting the raw coal.

The Washing of Coal by Gravity Separation. J. Griffiths. (Proceedings of the South Wales Institute of Engineers, 1935, vol. 51, pp. 199–233: Colliery Guardian, 1935, vol. 150, May 17, pp. 890–893; May 24, pp. 935–937). The washing of coal by the Chance process is outlined. The plant embodies the use of sand and water, the sand being kept in agitation so that it has the effect of increasing the specific gravity of the water and enables the lighter particles to float. Gravity separation takes place in a cone classifier, the fines having been previously extracted from the raw coal. Throughout the plant care is taken to minimise size degradation; it is stated to give very satisfactory results.

The Cleaning of Coal by Means of Dense Media. C. Berthelot. (Revue de l'Industrie Minérale, Mémoires, 1935, No. 349, July 1, pp. 299–316). The author presents a study of the theory and practice of coal-cleaning by means of dense media; the viscosity of these media, its variation and means for preventing it; and the physical stability of a dense medium composed of water, barytes and clay. He describes the plant at the Sophia-Jacoba Colliery, Holland, where the de Vooys process is in operation; the purification and stabilisation of the medium; the operating results, and the working costs. He concludes with a reference to the preparation of very pure coal and the separation of fusain, and the construction of a modern washery.

The Response of Coal-Cleaning Practice to the Demands of Modern Mining. G. Rau and F. F. Ridley. (Institution of Mining Engineers, July 11, 1935: Colliery Guardian, 1935, vol. 151, July 12, pp. 51–54). The development of mechanised

mining methods and the demand for improved quality in the coal produced, have necessitated special attention being paid to the methods of cleaning and grading coal. Wet cleaning, by jigs or troughs, is applicable in many cases, but if the raw coal is sufficiently dry, pneumatic cleaning may be suitable. If a high-grade product is required, or if the inherent ash is high, gravity separation should be adopted. Fines may be treated by flotation or by tabling. In wet-washing processes slurry formation is greatly reduced if the coal is subjected to a preliminary de-dusting operation. Wet washing of large coal and dry cleaning of fines appears to be the logical direction in which progress may be expected.

The Hoyois Washery at Rockingham Colliery. T. Sinclair. (Colliery Guardian, 1935, vol. 150, May 17, pp. 887–890). The washer is stated to be the first of its kind erected in Great Britain. It is designed for a capacity of 80 tons per hr. of slack, ½-in. to 0-in., the main units being a Lessing de-duster and the Hoyois washer for dealing with the de-dusted coal. The plant is linked with an existing washery for the treatment of 3-in. to ½-in. coal.

The New Preparation Plant at the Rising Sun Colliery. K. C. Appleyard. (Institution of Mining Engineers, July 11, 1935: Colliery Guardian, 1935, vol. 151, July 12, p. 59). The plant is designed to handle a throughput of 160 tons per hr. of 4-in. to 0-in. coal. The coal is sized on jigging screens, and the oversize (4-in. to 1-in.) passes to a washery, and is subsequently sized for the market. The undersize (1-in. to 0-in.) goes to a dry-cleaning plant and de-duster. To meet the demand for high-quality small coal (below ½-in.) this may be separately treated by washing, and the fines improved by flotation. A thermal drier is incorporated in order to reduce the moisture content of the cleaned coal, and blending equipment is also included.

Dry Cleaning at Silksworth Colliery. (Colliery Engineering, 1935, vol. 12, June, pp. 195–198, 209). The plant is designed to treat 130 tons of raw coal per hour, from 2-in. to zero size. The coal is screened on four Hummer units, giving progressive sizing. The sized material is treated by Super Vee machines, and the coal below $\frac{1}{32}$-in. is dealt with in fines separators. In order to reduce breakage, Redler conveyors have been incorporated.

New Screening and Dry-Cleaning Plant at Baddesley Colliery. (Colliery Engineering, 1935, vol. 12, Sept., pp. 289–292). **Dry Cleaning Plant at Baddesley Colliery.** (Iron and Coal Trades Review, 1935, vol. 131, Oct. 4, pp. 553–555). The chief features of the plant are the screens and belts which form the main items of equipment, the pneumatic dry-cleaning plant and the cyclones which serve them.

Coal Slurry Treatment with "Unifloc" Reagent. (Fuel Economist, 1935, vol. 10, June, p. 891). **Treatment of Coal**

Slurry. (Iron and Coal Trades Review, 1935, vol. 131, Aug. 16, pp. 233–235). "Unifloc" is a flocculating reagent for use on coal slurries and colloidal suspensions. It is in use at the Emlyn Anthracite Colliery, Ltd. In this plant from 8,000 to 10,000 gal. of water per hr. are clarified by flocculation, and from 35 to 40 tons of fine coal are recovered. The flocculating plant is stated to operate on a profitable basis and is efficient in use.

The Propagation of a Zone of Combustion in Powdered Coal. H. E. Newall. (Fuel in Science and Practice, 1935, vol. 14, June, pp. 160–167). Samples of coal dust were elutriated and the sizes of the particles in each fraction were measured microscopically, the specific surfaces being then calculated. These values were plotted against the times of propagation of a zone of combustion through the dusts, and against the amounts of oxygen reacting with the coals during 2 hr. at 200° C. The volume of oxygen absorbed increased with increasing specific surface, but was not directly proportional thereto. Up to a certain limit an increase in fineness augmented the rate of propagation of combustion. To determine the relative ease of oxidation the fineness of the dusts must be standardised. However, after a certain period of grinding the time of combustion assumes a value which is not altered by further subdivision of the particles. It would appear possible to measure the relative tendencies of coals to oxidation by the limiting time of propagation of a zone of combustion along a standard "train" of coal dust under 200 mesh. (*See* Journ. I. and S.I., 1935, No. I. p. 338).

Pulverised Coal as a Standard Industrial Fuel. H. D. Tollemache. (Fuel Economy Review, 1935, vol. 14, pp. 21–24). It is pointed out that although pulverised fuel firing has many advantages the necessity for installing grinding plant has hitherto militated against its use. This drawback no longer exists, since it is now possible to obtain powdered fuel in bulk.

Firing Furnaces with Pulverised Fuel. (Iron and Coal Trades Review, 1935, vol. 131, Sept. 13, p. 385 : Foundry Trade Journal, 1935, vol. 53, Sept. 19, pp. 208, 218). Pulverised fuel firing has been installed in the air furnaces at the Empire Works of the British Chilled Roll and Engineering Co., Ltd., Middlesbrough. So far, the system has been applied to a 12-ton furnace previously fired with solid fuel ; the furnace has been sufficiently successful to warrant the conversion of two others (of 20 and 10 tons capacity) to this type of firing.

The Importance of Electrofilters in the Working of Modern Pulverised Fuel Boilers. F. Rathert. (Periodic Review of the Metallgesellschaft, 1935, May, No. 10, pp. 17–22). The adoption of pulverised fuel firing in modern boiler installations has rendered it necessary to install a suitable form of apparatus for the removal of flue dust, since the problem of dust entrainment is much more

serious with this type of firing than with lump coal firing. The particles of dust are so small—much of it being below 20μ in diam. —that only electrostatic precipitation will give the required degree of cleanliness. It is profitable to blow the dust back into the furnace, where the remaining combustible matter is consumed and the ash converted into slag; furthermore, no expense on account of dust removal is incurred. An installation in operation at Papeteries de Genval is described, and data concerning a number of other plants are also furnished.

The Briquetting of Small Coal. J. B. M. Mason. (Sands, Clays and Minerals, 1935, vol. 2, June, pp. 110–119). The binders which are used in the production of coal briquettes are examined from the point of view of suitability, and the author goes on to describe the Tilmanstone briquette plant. The small coal is screened through $\frac{11}{16}$-in. mesh and mixed with crushed petroleum bitumen; from 3 to 10 per cent. of this binder is incorporated, and about $1\frac{1}{2}$ per cent. of an emulsion. The mixture goes to a disintegrator to be blended, and is conveyed to two cylindrical heaters, heated with steam at 450° F. It enters a pug mill and drops between two rolls which mould the mixture into ovoid briquettes. Rectangular briquettes are made in a press. The products must pass certain tests for hardness, shatter and free-burning ability in order that they may prove suitable for the domestic market.

COKE.

Investigations on Saar Coals from the Point of View of Raw Material and Coking Technique. H. Hoffmann and F. L. Kühlwein. (Glückauf, 1935, vol. 71, July 6, pp. 625–639; July 13, pp. 657–664). The authors present the results of an investigation of the petrographic constituents of Saar coals and their influence on the coking process.

The Pishel Rapid Test for Determining the Cokability of Coals. H. Romberg. (Glückauf, 1935, vol. 71, June 15, p. 569). In Pishel's proposed method, a small quantity of coal is ground in an agate mortar; on tipping it out again the degree of its adherence to the mortar is noted. The more strongly and closely it sticks to the mortar, the better is its cokability. The present author has tested the applicability of this method to German coals. He found, however, that, though in many cases there was a connection between the adherence of the powdered coal to the mortar and the cokability, yet there were a great many cases where this was not so, and he concludes that the test is unsuitable for German coals.

The Reaction of Bromine with a Coking Coal, Including a Study of its Unsaturation. J. F. Weiler. (Fuel in Science and Practice, 1935, vol. 14, July, pp. 190–196). The rate of reaction of a coking coal with bromine in carbon tetrachloride solution is shown to be practically independent of particle size (between 1μ and 100 mesh) and almost the same for dried and undried samples. Also, it was nearly the same for the residue from the petroleum ether insoluble fraction of a benzene pressure extraction as for the original coal. Reaction occurred by both substitution and addition, the former predominating. Approximately two-thirds of the bromine combining with coal was hydrolysed by alcoholic potassium hydroxide solution and was therefore linked to aliphatic carbon atoms.

Carbonising Properties and Constitution of Alma Bed Coal from Spruce River No. 4 Mine, Boone County, W. Va. A. C. Fieldner and others. (United States Bureau of Mines, 1935, Technical Paper No. 562).

A Coke-Oven Reorganisation in Durham. (Coal Carbonisation, Supplement to Colliery Engineering, 1935, June, pp. 10–15). The Mid-Durham Carbonisation Co., Ltd., has erected a new battery of 50 Still ovens and adapted its by-product plant to meet the increased yield from an additional 150 tons throughput per day. The alterations made and new items of equipment installed are described.

The Improvement of the Yield of Benzol from Coke-Ovens. F. Rosendahl. (Montanistische Rundschau, 1935, vol. 27, May 16, pp. 15–16). A brief review of a number of processes devised for increasing the benzol yield from coke-ovens.

Influence of Interior and Roof Evacuation on the Yield of Coking Products. W. Litterscheidt and W. Reerink. (Glückauf, 1935, vol. 71, May 18, pp. 461–470). The authors describe tests made on the Still process of interior evacuation of the distillation products from the coke-oven and the Goldschmidt and Tillmann process of roof evacuation, to determine to what extent these methods improve the yield of products. The coking coal used contained 18·2 per cent. of volatile matter. With interior evacuation, the calorific value of the gas was slightly lower, which was balanced by a certain economy of heating gas. The composition of the gas was not markedly different from that of gas evacuated in the normal manner; the volume and composition of the "interior" gas was closely dependent on the negative pressure, measurable quantities being obtained only with a suction of 20 mm. water gauge. The yields of crude benzol and light oils in the tar were increased; this increase was due to the presence of greater quantities of low- and high-boiling fractions, for the yield of pure benzene was lower. The tar produced was also increased; it contained a particularly high proportion of

constituents boiling at up to 200° C., and much less pitch than does normally produced tar. Less ammonia was produced ; the amount and condition of the coke were unchanged, though the coke abrasion seemed to be less. The coking period was shortened. The tests results obtained with roof evacuation did not differ much from those obtained under normal conditions. The deviations were not all in the same direction, and they may possibly be due to unavoidable experimental differences.

The Production in the Coke-Oven of Oils and Tars from Coal. G. E. Foxwell. (Gas World, 1935, vol. 103, Sept. 7, Coking Section, pp. 9–12). The author shows that the results of laboratory tests on the oil and tar yield from coals points to a process of continuous chemical reaction. He then examines the effect of the path of travel of the volatile matter evolved, and in this connection discusses the influence of devices such as the Goldschmidt top channel on the output of benzol and other volatiles.

Increasing the Yield of Light Spirit—I. A. Jenkner. (Coal Carbonisation, Supplement to Colliery Engineering, 1935, vol. 1, June, pp. 6–9). In this article it is shown that the conditions of distillation of coal—and particularly the path taken by the gases—have a considerable influence on the yield of light spirit. The advantages of top flues of the Goldschmidt type in this connection are stressed, and an account of an investigation on a twin-flue Otto oven, in which the gas travel was shortened by building in an iron pipe, is furnished. This gave a decreased yield of benzol ; experiments were therefore undertaken so as to introduce additional cracking by means of an electrically-heated gas flue. Full-scale experiments indicated an improved yield.

Increasing the Yield of Light Spirit—II. A. Jenkner and L. Nettlenbusch. (Coal Carbonisation, Supplement to Colliery Engineering, 1935, vol. 1, July, pp. 27–30). In this article the influence of additional cracking on the yield of benzol is considered ; a laboratory apparatus to investigate the effect of additional cracking on the yield of by-products is described, and some results obtained by its use are reproduced. In conclusion, it is pointed out that, owing to the varying carbonising conditions which always exist in practice, the benzol yield is not always increased by building top flues into the ovens. The most favourable conditions are attained when the top flues can be heated independently of the oven chambers. Additional cracking of the gases from the low-temperature carbonisation of coal leads to a reduction of both light spirit and tar yields, accompanied by a marked increase in the content of heavy hydrocarbons in the low-temperature gas.

Basic Data on the Question of Roof Evacuation. H. Kreuger, B. Hofmeister, and E. Krebs. (Glückauf, 1935, vol. 71, Mar. 9, pp. 221–230). At many coke-oven plants nowadays, the gas

is evacuated from the coke-ovens through a flue built into the roof, and increased benzol yields are obtained. Previously it has been believed that this increased yield was due to the protection of the gas from thermal effects likely to cause breakdown ; calculation of the time of sojourn of the gas in the oven shows, however, that roof evacuation does not shorten the stay of the gas in the oven, but that through it the gas undergoes after-treatment at cracking temperatures. Laboratory tests show that this thermal after-treatment can increase the benzol yield, but its effect is largely dependent on the extent to which reactions due to heat have proceeded in the oven ; the ultimate yield of benzol represents the balance of concurrently proceeding reactions of formation and breakdown. Further tests show what changes may be undergone by benzol, toluol and xylol in the coke-oven atmospheres at temperatures between 500° and 1,000° C. ; the substances produced are described, and it is demonstrated that at the usual temperatures no decomposition occurs in the gas-collecting space. An oil derived from an " interior " tar from a Still oven was submitted to the same treatment as benzol and its homologues under conditions similar to those reigning in the roof flue, and the formation of low-boiling-point hydrocarbons was observed ; their nature and amounts are described. The path of the gas in the oven is discussed, and it is shown on the basis of the results obtained that the increased benzol yield is explicable by the cracking theory. The authors conclude by describing an effect of the roof flue not previously observed. In an oven of the ordinary type, particularly with short coking periods, considerable gas pressure differences occur which lead to the passage of the raw gas into the heating flues. By evacuating the gas through a roof flue these pressure differences are largely avoided, so preventing the passage of the gas into the heating flues, and the consequent loss of products.

Gas Suction at the Crown of Coke-Ovens. (Iron and Coal Trades Review, 1935, vol. 130, May 31, pp. 945–946). An English abstract of the foregoing article by Kreuger, Hofmeister, and Krebs.

New Adaptations of Old Distillation Processes. (Coal Carbonisation, Supplement to Colliery Engineering, 1935, June, pp. 3–5). The various attempts made to recycle the distillates are considered, and the main features of the Knowles process for coking blended and feebly coking coals are examined.

Development of Low Temperature Carbonisation during 1934. F. S. Sinnatt. (Fuel Economy Review, 1935, vol. 14, pp. 3–4). The progress of low-temperature carbonisation during 1934 is briefly surveyed.

Combined Low-Temperature Carbonisation and Combustion. D. Brownlie. (Steam Engineer, 1935, vol. 4, Aug., pp. 465–468).

An account of experimental work in connection with combined carbonisation and combustion plants now being carried out by J. S. Morgan at Roderidge Hall, near Wingate, County Durham, is given. Two processes have been evolved. In the first—which is primarily intended for the operation of large water-tube boiler plants provided with travelling grate stokers—the stokers of one boiler are run at a much higher speed than normal in order to deliver incandescent coke at the end instead of ash and clinker. This red-hot material is mixed with raw coal in a rotary mixer, so as to give very rapid low-temperature carbonisation, in about 15 min., say, the gases and vapours being discharged to a by-products plant. The coked fuel is burnt on the travelling grates of adjoining stokers, together with the residual gas. The second process is intended for smaller industrial steam generation plants. A charge of coal is introduced into a container rather like a concrete mixer but lined with firebrick. The coal is ignited and a blast of air is admitted at the base to gasify (partially) the coal. Another charge of raw coal is added and the container is rapidly rotated, the raw and red-hot coal being thoroughly mixed. Rapid carbonisation takes place, the gases evolved being led away to a recovery plant. The charge is " blown " again, and the coked produced is discharged periodically to be used under steam boilers.

Low-Temperature Carbonisation Ovens. Fuels and Motor Fuels. E. Marcotte. (Arts et Métiers, 1935, vol. 87, May, pp. 105–109; June, pp. 124–128). After a general discussion of various matters related to the low-temperature carbonisation of coal, the author deals with the technique of the process and reviews a number of types of plant.

The " Suncole " Low-Temperature Process. (Coal Carbonisation, Supplement to Colliery Engineering, 1935, vol. 1, July, pp. 38-40, 43). Particulars are given of the scheme developed by British Coal Distillation Ltd., for producing domestic fuel from non-coking duff. The coal is carbonised in a retort consisting of a rotating inclined steel drum, containing steel tubes through which the coal travels, and heated by hot inert gases at 650° C., which pass both inside and outside the tubes. The coking time is 2½ hr. and the hot residual coke is conveyed to a heat exchanger where its sensible heat is yielded to wet, washed coke. The coke is washed, dried and briquetted with pitch, the briquettes being stoved at 550° C. for 40 min. to harden them and drive off the volatiles. The plant is stated to be capable of operating on a profitable basis.

The Electrical Distillation of Coal. C. Ficai. (Chimie et Industrie, 1935, vol. 34, July, pp. 22–28). The author first refers briefly to the few attempts that have been made to distil coal electrically, and then gives an account of his own experiences in this field. The advantages which this process may possess depend

essentially on the local conditions, so that in certain cases the method of distillation might supplement that employed in a gas works, but could never replace it.

Removal of Benzol from Coke-Oven Gas by Refrigeration. (Coal Carbonisation, Supplement to Colliery Engineering, 1935, vol. 1, Sept., pp. 72–74). This is an article translated from "L'Industrie Chimique Belge." Refrigeration is an attractive method of removing benzol from coke-oven gas. It is accomplished by combined pressure and cold; the stages are: (a) Compression of the benzol-bearing gas; (*b*) primary cooling; (*c*) further cooling by heat exchange; (*d*) final refrigeration; (*e*) expansion of the benzol-free gas. In the first stage the gas is compressed to about 60 lb. per sq. in. (or less) by means of turbo-compressors; in the second stage the gas is cooled to about 20° C. by circulating it through water-cooled tubes; in the two succeeding stages the gas is reduced to – 20° C. by heat exchangers, the temperature being further reduced to – 40° C. by ammonia refrigerators. The benzol-free gas flows through the heat exchangers and cools the incoming benzol-bearing gas; it is then allowed to expand. A recovery of 92 to 94 per cent. is claimed, and if liquid ammonia is already available this may be used as the refrigerating agent.

The Ternary System Sulphuric-Acid/Ammonium-Sulphate/Water.—Application to the Recovery of Ammonia in Coke-Oven Plants. P. Locuty. (Revue de l'Industrie Minérale, Mémoires, 1935, No. 348, June 15, pp. 275–297). The author presents a study of the ternary system $(NH_4)_2SO_4/H_2SO_4/H_2O$ in the temperature interval 10° to 90° C. Various acid salts have been examined and their limits of stability determined. Evidence has been found for a new salt, $4(NH_4)_2SO_4 . H_2SO_4$, and a zone of solid solutions of $(NH_4)_2SO_4$ with either $3(NH_4)_2SO_4 . H_2SO_4$ or $4(NH_4)_2SO_4 . H_2SO_4$. From this study emerge the following points regarding the recovery of ammonium sulphate in coke-oven plants: (*a*) The production of neutral sulphate becomes easier as the temperature is raised; (*b*) the maximum content of H_2SO_4 in the saturated solution from which neutral sulphate can be obtained varies with the temperature; at 75° C., the usual working temperature, it amounts to 18 per cent.

Investigations to Explain the Process of the Thickening of Wash Oil. F. Schulte. (Glückauf, 1935, vol. 71, July 13, pp. 653–657). From his experiments on the thickening of wash oil, the author concludes that sulphuretted hydrogen has the greatest effect on wash oil at high temperatures; next comes oxygen, and its influence is stimulated by the catalytic action of iron and iron compounds. Phenol favours the thickening process. At raised temperatures, even in nitrogen atmospheres, wash oil is not completely stable. Continuous regeneration of the same

wash oil considerably increases the period of utility in a given plant. The regenerated wash oil has an increased power of absorbing benzol, so that the efficiency of washing is improved.

The Continuous Regeneration of Benzol Wash Oil. A. Thau. (Gas World, 1935, vol. 103, Aug. 3, Coking Section, pp. 12–14). It is pointed out that in the operation of a benzol plant thickening of the wash-oil is almost inevitable ; it is necessary, therefore, to adopt some method of renewal or regeneration. In the Weindel process continuous distillation is employed. The distillation is effected under vacuum at a reduced temperature, and gas introduced into the still under pressure serves to agitate the boiling wash-oil. The residue formed has the characteristics of a normal briquetting pitch. The regenerated wash-oil is returned to the benzole plant. The use of this regenerating plant is said to lead to a higher yield of benzole and a lower consumption of steam.

The Reactivity of Coke. A. Reis. (Chimie et Industrie, 1935, vol. 33, June, pp. 1297–1306). The part played by " reactivity " in the various uses to which coke is put has been studied by many investigators, but the results obtained do not agree. Definitions of reactivity are based on the speed of the reaction of the coke with either carbon dioxide or air at raised temperatures under prescribed conditions, and numerous methods have been devised for measuring the reactivity in small laboratory furnaces. There are several processes capable of characterising the types of cokes, but they give far from comparable results among themselves. The problem of coke reactivity is very complex : it depends both on the chemical character of the coke and on its porous structure. By modifying the previous treatment of the coal or the mode of operation of the coke-oven, the reactivity of the resulting coke can be raised or lowered considerably, and in this way the production of cokes of well-defined qualities, well suited to the uses for which they are required, can be assured.

Coking Coals for Carbonisation. J. Roberts. (Coal Carbonisation, Supplement to Colliery Engineering, 1935, vol. 1, Sept., pp. 79–81, 84). The author attributes the strength and high shatter index of metallurgical coke partly to the deposition of graphitic carbon on the cell walls, but mainly to the thickness of the cell walls. Thick walls tend to produce lumpy rather than columnar pieces of coke ; the melting and intermixing points of the coals also influence the physical character of the coke, while blending has quite a powerful effect.

The Preparation, Marketing and Utilisation of Coke. W. L. Boon. (Institution of Gas Engineers, June 6, 1935 : Gas World, 1935, vol. 102, June 1, pp. 585–588). The author deals with quenching, dry-cooling, conveying, cutting and screening, breeze

production, storage and grading in connection with gasworks coke. Marketing and utilisation are also discussed.

Transhipment of Coke in Storage Yards. (Demag News, 1935, vol. 9, May, p. 31). The handling of coke during transfer from the storage yard, where the material is generally stored as large dumps, requires the use of an efficient grab. The " Polyp " grab, consisting of eight shells which readily dig into the coke, is stated to be an efficient form of coke grab for this type of work. It is slung from a jib and may be mounted on a rail travelling gear or on a caterpillar track.

GASEOUS FUEL.

Natural Gas in Steel Plants. K. Emmerling. (Iron and Steel Engineer, 1935, vol. 12, June, pp. 337–341). The author is of the opinion that although natural gas may be more expensive in first cost there are many circumstances in which its use may prove economical. The application of natural gas to the firing of open-hearth furnaces is dealt with; reference is made to the design of gas ports and the control of fuel input to the furnace, and certain principles essential to the proper operation of natural-gas-fired open-hearth furnaces are enumerated. It is shown that natural gas can readily be applied to the heating of soaking pits and reheating furnaces if only suitable designs are adopted. The use of luminous-flame burners of small size and the introduction of positive furnace-atmosphere control have enabled this fuel to be applied to annealing furnaces most successfully.

New Gas Producer Plants for Coke-Ovens. (Demag News, 1935, vol. 9, May, pp. 24–31). The development of long-distance gas transmission has left certain producers with a surplus of coke. The breeze and coke waste may be economically gasified and the resultant gas used instead of coke-oven gas for heating the coke-ovens, thus effecting a saving in the consumption of high-grade coke-oven gas. In this article such a gas producer installation is described. The producers are of the revolving grate, water-sealed, water-cooled, mechanically fed type. The gas issues at 600° C. and is cleaned and cooled before passing to the coke-ovens, waste-heat boilers being used to extract the sensible heat. Electric grabs deliver the coke to the producers. The results obtained have fully realised the expectations of the designers.

Theoretical Study of Gasification. P. Coste. ' (Revue de Métallurgie, Mémcires, 1935, vol. 32, May, pp. 225–229). The author presents a theoretical discussion of the reactions which take place in gas producers.

The De-Dusting of Cold Gases. L. Lahoussay. (Revue de l'Industrie Minérale, Mémoires, 1935, No. 350, July 15, pp. 343–

353). A general discussion of the methods employed for de-dusting "cold gases," that is, gases at temperatures below 100° C., a problem encountered in plants such as coal screening or cleaning plants, drying ovens, crushers, &c., where this must be done before releasing the gas or air into the atmosphere or before using it again. Whereas the waste gases from furnaces generally contain dust of no commercial value, in the present case the recovery of the coal dust is of importance. Wet, dry and electric methods of de-dusting are considered.

Electrostatic Precipitation for Cleaning Industrial Gases. H. W. Wagner. (Fuel Economist, 1935, vol. 10, June, pp. 895–899; July, pp. 942–945; Aug., pp. 971–973). Electrostatic precipitators make use of the ionisation of the gas brought about by a high-tension discharge. The dust particles become ionised dispersoids and are propelled towards the receiving electrode by electrostatic force. It is possible in this way to secure the precipitation of very fine dust with efficiencies of almost 100 per cent. Commercial precipitators are of the plate or tube type. The discharge electrodes are of fine wire, tensioned by weights and prevented from oscillating by being clamped together at the free ends. The wire should be strong enough to withstand the mechanical stresses imposed, but fine enough not to require excessive operating voltages. Square wires have been found to be very effective. The dust is released from the electrodes by rappers. The electrical equipment consists of a transformer and rectifier; the latter may be of four types: (1) the mechanical rectifier, (2) the valve, (3) the d.c. generator, (4) the metal rectifier. The author describes some typical installations.

Gas Cleaning and Sludge Recovery. F. Wille and A. Mohr, jun. (Blast Furnace and Steel Plant, 1935, vol. 23, Aug., pp. 529–532). **New Gas Cleaning System Uses Hot Washing Stage.** F. Wille and A. Mohr, jun. (Steel, 1935, vol. 96, May, 13, pp. 42–43, 56). The Simplex gas cleaning and sludge recovery system is described. The gas is cleaned in two stages: (1) The hot washing, humidifying and conditioning stage; (2) the condensing and cooling stage. The sludge formed during the hot stage is removed by means of a diaphragm pump. In stage (1) the gas is rough cleaned by hurdles and settling pans, and is passed through a Bassler rotor. In stage (2) the gas is cooled, whereupon the moisture condenses on the fine dust and so precipitates it. The condensing and cooling equipment consists of two banks of static hurdles and a Bassler rotor.

Some Experiences in the Purification of Coke-Oven Gas. J. R. Marshall. (Coke-Oven Managers' Association, May 3, 1935 : Gas World, 1935, vol. 102, June 1, Coking Section, pp. 58–60: Iron and Coal Trades Review, 1935, vol. 130, June 28, p. 1095). An account is furnished of some practical difficulties

met with in removing H_2S from coke-oven gas, using iron oxide as a purifier.

The Klönne Waterless Gasholder. W. A. Harper. (Wales and Monmouthshire Association of Gas Engineers and Managers, July 8, 1935 : Iron and Coal Trades Review, 1935, vol. 131, July 26, p. 125 : Gas World, 1935, vol. 103, July 27, pp. 70–72). The advantages of this type of holder over the water-seal type are stated to be : Lighter weight and lighter foundations ; smaller ground area ; an almost uniform gas pressure is maintained ; the gas is in a dry condition ; the maintenance required is less. The shell of the holder is constructed of $\frac{3}{4}$-in. steel sheets riveted to a frame of joists and columns. Within the shell is a sliding piston with an oil-filled cup at its periphery, into which a flexible sheet-steel ring carrying a sealing ring is allowed to dip. The seal is composed of six rings of rubber and cotton, 4 in. by 1 in. in section ; it is pressed against the interior of the holder by weighted levers. In an alternative design the oil-cup is dispensed with and a specially constructed fabric is used. The Klönne holder is stated to be completely gas-tight in operation. Hard-wood rollers are used to guide the piston. Friction between the shell and the sealing ring is reduced by a thin layer of grease supplied by means of loaded lubricators.

LIQUID FUEL.

The Scottish Shale Oil Industry. (Chemical Age, 1935, vol. 33, July 13, pp. 31–34). The technique of shale oil production at a typical Scottish plant (Deans Oil Works) is described. The shale is distilled in vertical gas-fired retorts of the Pumpherston type, arranged for continuous operation, a maximum temperature of 1,300° F. being attained. Shale oil is evolved at 650° to 900° F., and at higher temperatures the steam which is allowed to pass into the retort converts the nitrogen in the shale into ammonia and the fixed carbon into water-gas. The products are drawn off by exhausters and the crude oil and naphtha are subsequently refined, giving rise to several grades of naphtha and shale oil, and paraffin wax and coke.

Hydrogenation. F. S. Sinnatt and J. G. King. (Coke-Oven Managers' Association, Apr. 5, 1935 : Colliery Guardian, 1935, vol. 150, May 31, pp. 982–984). The authors discuss the effect of temperature and of catalysts on the hydrogenation of coal, and pay some attention to the hydrogenation of tar.

Research on the Hydrogenation of a Colloidal Solution of Coal. J. M. Pertierra. (Journal of the Institute of Fuel, 1935, vol. 9, Oct., pp. 16–23).

Combustion Properties of Industrial Propane. W. Z. Friend. (Heat Treating and Forging, 1935, vol. 21, June, pp. 297–301). Propane is the principal liquefied petroleum gas commercially available; it may be liquefied by moderate pressure, and allowed to vaporise when required for use. In this article useful data on the properties of pure and industrial propane and on the air required for combustion, the calorific value, &c., are summarised.

Combustion Properties of Industrial Propane. W. Z. Friend. (Heat Treating and Forging, 1935, vol. 21, July, pp. 345–349; Aug., pp. 394–396). The author compares the rates of flame propagation, flame temperatures, limits of flammability, ignition temperatures, available heats, specific heats and types of flame obtained with propane and other gaseous fuels. Among the matters dealt with are the production of various types of furnace atmosphere, the use of propane in refrigeration, and the vapour pressure and dew point curves. Finally, a table of the thermodynamic properties of propane is reproduced, and the advantages of propane as an industrial fuel are summarised.

PRODUCTION OF IRON.

PIG IRON.

Blast-Furnace Plant of the South African Iron and Steel Industrial Corporation at Pretoria. (Iron and Coal Trades Review, 1935, vol. 130, May 24, pp. 890–891). Two iron ores are used ; a siliceous hematite containing about 48 per cent. of iron and up to 21 per cent. of silica, and a very pure hematite averaging 67 to 68 per cent. of iron. The blast-furnace has an output of about 500 tons per day, and is designed on orthodox lines. The site is some 4,500 ft. above sea-level, and the low barometric pressure necessitates the use of generous turbo-blower equipment to meet this condition ; care is also taken to conserve water supplies. The gas is cleaned by a Lodge-Cottrell electrostatic cleaner and waterless gas-holders are installed to maintain a reserve of gas. The pig iron is taken direct to the steelworks in 50-ton ladles ; any excess is cast in a pig-casting machine. The plant is stated to operate satisfactorily.

Blast-Furnace Practice in Belgium and Luxemburg. G. B. Butler. (Proceedings of the Cleveland Institution of Engineers, 1934–35, No. 3, pp. 227–291). A general outline is given of blast-furnace construction and practice in these countries.

A Review of Blast-Furnace Developments in the U.S.S.R. B. M. Suslov. (Iron Age, 1935, vol. 136, Aug. 1, pp. 12–17). It is shown that during the last five years great improvements have been made in blast-furnace practice in Soviet Russia. Many furnaces now deliver a daily output of over 1,000 tons of pig iron, equal to the best figures for American and European practice. The size of the furnaces has been gradually increased, and the coke consumption per ton of pig iron reduced. Difficulties have been met with—such as poor coke quality, and fineness of ore—but these are being overcome. Russia is now the second largest producer of pig iron in the world. The author furnishes statistics concerning pig iron output, the size of furnaces, &c.

Development of the Brosius Type of Tap-Hole Clay Gun. A. Wapenhensch. (Stahl und Eisen, 1935, vol. 55, July 25, pp. 809–810).

Operation of Blast-Furnace Blowers. (Iron and Coal Trades Review, 1935, vol. 131, July 5, pp. 6–7). Illustrated particulars are given of the Scherbius system of regenerative control for blast-furnace blowing plants.

Acceptance Tests on an Alsthom Turbo-Blower. M. Steffes and R. Welter. (Chaleur et Industrie, 1935, vol. 16, Apr., pp. 161–178). The authors record, in graphs and tables, the results obtained in making acceptance tests on a turbo-blower for the blast-furnaces of the Audin works of the Société Minière des Terres Rouges. The turbine, of the Zoelly type, was built by the Société Alsthom, of Belfort; the blower, built by the Société Alsacienne de Constructions Mécaniques, Mulhouse, is of the Pokorny type, and is direct-coupled to the turbine.

Analysis of Design and Construction of Hot Blast Stoves. A. Mohr, jun., and F. Wille. (Iron Age, 1935, vol. 136, July 4, pp. 12–17). A hot-blast stove should possess the greatest possible rate of heat transfer compatible with adequate structural stability and simplicity of design. The authors analyse the factors which bear upon the hot-blast stove efficiency, with special reference to the Simplex type of stove. They are of the opinion that no substantial gain in heat transfer efficiency is to be obtained by reducing the thickness of the brick below 2 in., whilst structural stability demands the use of checker work sufficiently strong to withstand contraction and other stresses. Data dealing with heat transfer, turbulent flow and the effect of staggering the brickwork are provided, and the influence of zoning on stove efficiency is discussed.

Brickwork for Blast-Furnace Stoves. (Iron and Coal Trades Review, 1935, vol. 131, Sept. 13, pp. 391–392). In order to obtain the greatest heat transfer efficiency it is suggested that the top section, near the dome, should have a high mass/surface ratio in order to take advantage of heat radiation, whilst succeeding zones should offer sufficient resistance to the gas flow to promote turbulence. The bricks should also be strong, cheap and easy to instal. The Hotspur vaned checker system is described, in which the upper zone contains rectangular bricks provided with a blunt cruciform boss and the lower zones are filled with bricks fashioned with vanes or projections. It is claimed that this method of construction takes full advantage of heat transfer by both radiation and convection.

The Use of Oxygen-Enriched Blast in Blast-Furnace Operation. W. Lennings. (Stahl und Eisen, 1935, vol. 55, May 16, pp. 533–544; May 23, pp. 565–570). For a given coke consumption in the blast-furnace, the heat available at the maximum temperature for the hearth reactions is raised by the addition of oxygen to the blast, that is, a given heat requirement can be covered by less coke in the charge. By the reduction of the volume of the furnace gases, the amount of heat supplied to the shaft is reduced; the temperature and heat losses at the throat decrease, while the total amount of heat carried into the furnace in the reduced volume of blast becomes lower. In the region of maximum

temperature in the furnace, the temperature differences between the gas and the charge increase, and consequently the speed of smelting also. Tests carried out by the Gutehoffnungshütte in an experimental furnace with an output of 50 to 60 tons per day showed that in the production of basic Bessemer pig, stahleisen, spiegeleisen and foundry iron with a coke consumption of 1,000 to 1,150 kg., by the use of oxygen-enriched blast at from 700° to 800°, a saving of 100 to 150 kg. of coke per ton of pig can be effected. For materials requiring higher coke consumptions, such as ferro-manganese, the economy of coke amounts to from 300 to 400 kg. Linings and cooling appliances, even in the presence of highly basic slags, have the same durability during operation with oxygen-enriched blast as with normal air blast. The extent to which the blast can be enriched with oxygen is limited by the extent of the coke consumption and the throat gas temperature. In general, enrichment up to about 24 to 26 per cent. is suitable, and up to 30 per cent. for ferro-manganese; enrichment above these figures may lead to a deficiency of heat in the shaft. The higher the throat temperature and the coke consumption, the more advantageous is the enrichment of the blast; with low throat-gas temperatures, as in the smelting of minette ore, the process is not applicable. For the production of basic Bessemer pig, for which the coke consumption is low, the application of oxygen enrichment is restricted to special cases. Highly heated blast (which is, therefore, expensive) can be replaced by oxygen-enriched blast preheated to a less extent, and in this way an economy is effected. In furnaces which work with difficulty and which have a tendency to hanging, oxygen-enriched blast accelerates the rate of working. The high hearth temperatures attainable by the use of oxygen-enriched blast lead to an increased reduction of silicon with a reduced coke consumption; further, slags richer in lime can be melted than in normal operation with air. In this way, the manganese content of blast-furnace slags can be considerably reduced and a larger proportion of the manganese turned to account. The experiments show that with oxygen-enriched blast the output of the furnace can be increased without raising the resistance of the charge. An improvement of the quality of the pig iron, apart from its lower sulphur content, was not observed. The economy of the process depends upon local conditions; it is determined by the cost of production of the oxygen and also by the cost of the coke. For the cost conditions in the Ruhr district and a cost of 7 *RM.* per 1,000 cu. m. of 80 per cent. oxygen, the experiments on high-silicon pig irons showed a coke economy amounting to about 1 to 1·5 *RM.* per ton of pig. In some cases there is a further economy due to the increased turning to account of manganese and to the increase of the furnace output. In the

production of ferro-manganese the process is advantageous even with coke at the lowest possible price. Experiments on the production of blast-furnace Portland cement slags simultaneously with the pig iron in the blast-furnace with oxygen-enriched blast were technically successful, but at the present state of the experiments the process cannot yet be applied to the operation of large blast-furnaces.

New Processes for the Production of Oxygen for the Enrichment of Blast-Furnace Blast. E. Karwat. (Stahl und Eisen, 1935, vol. 55, Aug. 8, pp. 860–863). The author describes the Linde-Fränkl process of making oxygen for the enrichment of blast-furnace blast. The essential recent improvement in this process consists in the replacement of the earlier tubular recuperators used in the fractionation of the liquified air by frequently-reversed metal-filled regenerators. In a plant capable of an output of 4,430 cu. m. of 80 per cent. oxygen (sufficient for an ironworks making 570 tons of manganiferous pig iron [*Stahleisen*] per day), the cost amounts to 6·72 *RM.* per 1,000 cu. m., with a gas-driven compressor and 10·70 *RM.* with an electrically-driven turbo-blower and high-pressure compressor.

The Physico-Chemical Principles of the Burdening of Iron Ores (Part III.). J. Klarding. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Sept., pp. 127–129). Reduction experiments were carried out on mixtures of iron ore, alumina and silica. It appears that the compound silimanite, formed from the alumina and silica, is not so stable as the chemical compounds which may be formed from the various iron oxides and the other two oxides. Alumina and silica react with ferrous oxide in sequence. For burdening in the usual sense, the experimental results indicate that, in the presence of silica and alumina, the optimum burden proportion corresponds to the composition $3CaO.SiO_2.Al_2O_3$. (*See* Journ. I. and S.I., 1935, No. I. p. 354).

Silica and Alumina in Iron Ores. T. L. Joseph and W. F. Holbrook. (Blast-Furnace and Steel Plant, 1935, vol. 23, May, pp. 313–316). Ore shipped from the Lake Superior district has tended, within recent years, to carry a large proportion of concentrates. The concentrates average about 16 times as much silica as alumina compared to a corresponding ratio of about 3·7 for directly mined ore. Hence the ore treated in the blast-furnace now carries higher silica/alumina ratios. The authors discuss the effect of alumina on the desulphurising action of blast-furnace slag. They regard alumina as an acid constituent, and express the basicity of the slags investigated in terms of the ratio $(CaO + MgO)/(SiO_2 + Al_2O_3)$. It is shown that desulphurisation improves as the alumina increases up to an optimum amount which varies with the basicity of the slag. This applies to slags in which alumina is substituted for silica; if alumina is regarded

merely as a slag diluent the improvement may be attributed to the increase in basicity. Tests were also carried out on six slags of differing basicity within the approximate temperature range 1,450° – 1,550°. Except in the highly basic slag, the substitution of alumina for silica improved the desulphurisation throughout this range.

The Equilibrium between Oxygen and Phosphorus in Liquid Iron Melts. H. Wentrup, K. Schwindt, and G. Hieber. (Druckschrift der Gesellschaft von Freunden der Technischen Hochschule Berlin über das Geschäftsjahr 1934, pp. 53–63 : Stahl und Eisen, 1935, vol. 55, Oct. 3, pp. 1068–1069). The reactions between oxygen and phosphorus in liquid iron melts have been investigated frequently before, but with contradictory results ; the present authors have repeated the examination of the equilibrium at 1,400° C., paying particular attention to the following experimental conditions : A neutral furnace atmosphere, impeccable crucible material, exact temperature measurement, sufficient melting time for the attainment of equilibrium, rapid quenching of the melts (water is unsuitable for quenching iron phosphate slags, because they absorb it), pure raw materials, and the analysis of the slags and metal after the test—the last point is most important because iron phosphates volatilise appreciably at the temperatures in question. The authors' results are plotted diagrammatically ; the results of other workers are also included and the curves are compared and the implications discussed. The difficulties in the way of calculating the equilibrium constants are also dealt with. Besides using five equations that have been suggested, the authors worked out the concentrations as molar percentages (per litre) and plotted them ; the results agree best with a hyperbolic curve based on the existence of Fe_3P in the metal and $3FeO.P_2O_5$ in the slag.

The Reducibility of Dwight-Lloyd Sinter and its Adjustment to the Ore Burden. K. Grethe and J. Stoecker. (Stahl und Eisen, 1935, vol. 55, June 13, pp. 641–647). The production of uniform sinter depends on the chemical behaviour of the raw materials and on the permeability to gas of the mixture. Iron-bearing substances in which the difference between the temperature of roasting and of melting is great are beneficial to the process. The gas permeability is much influenced by the size and shape of the particles, particularly of those under 0·1 mm. in size, and, also, by the binding power of the mixture. Regulation of the rate of travel of the roasting conveyor after the suction has been applied, and also of the moisture content, have proved advantageous. The reducibility of a given ore up to 1,000° C. usually does not vary. The reducibility of sinter is practically independent of the proportion of iron-bearing material contained in it. With higher proportions of fuel in the material to be sintered the slagging

of the sinter and its strength increase, while the index of reduction falls, with simultaneous commencement of decomposition of oxygen on the conveyor. The sum of the oxygen reduced on the conveyor and in the blast-furnace at up to 1,000° C. is practically invariable. For blast-furnace operation the porosity of the sinter is of the utmost importance; by carrying out the sintering in a suitable manner the sinter can always be adapted to the burden. The tests described in the paper are recorded in graphs.

WROUGHT IRON.

Wrought Iron in To-day's Industrial Picture. J. Aston. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 632). The subject is treated under three headings: (1) Manufacture, (2) characteristics related to the quality of the material, (3) adaptability and exploitation of product. The history of the puddling process is briefly outlined, and the Aston-Byers method—in which pig iron is first refined by Bessemerising, and then poured into an iron-silicate slag—is also described. This is followed by a consideration of the characteristics—including mechanical properties, chemical analysis, and microstructure—peculiar to wrought iron. The features which enable good quality wrought iron to be distinguished from poor or adulterated material are referred to. Finally the adaptability of wrought iron is noted.

DIRECT PROCESSES.

Present Status of Direct Production of Iron and Steel from Ores. R. S. Dean. (Mining and Metallurgy, 1935, vol. 16, pp. 185–186). According to the author sponge-iron processes should not be used for treating ore of blast-furnace grade, but should be confined to ores either of high-purity or from which concentrates of high purity can be obtained. The Bureau of Mines has developed a process of gaseous reduction in which natural gas is converted to carbon monoxide and hydrogen by passing it over nickel or alumina. The heat from this partial oxidation supplies that necessary for the reduction of preheated ferric oxide to iron. Sponge iron produced from high-purity ore can be converted to wrought iron by simple briquetting and forging at about 1,400°. A slag-metal mixture corresponding to a wrought-iron ball can be made by briquetting a mixture of fine ore and coal and heating at a proper rate at 1,450–1,500°. The "puddled ball" so formed can be squeezed in the usual manner to remove the slag.

Proposal for the Production of Crude Steel in the Blast-Furnace and for the Smelting of Acid Ores by a Duplex Process. M. Fränkl. (Stahl und Eisen, 1935, vol. 55, July 25, pp. 805–809). The author describes a method for the production from ore, in the blast-furnace, of low-carbon, low-silicon crude steel, which is then worked up into mild steel by treatment in a rotary refining furnace with ore reduced to the ferrous-oxide condition. Two structural forms for the blast-furnace have been developed. In the first, lump ore is fed into the central part of the furnace, with coke, limestone and fine ore round the circumference. To obtain better reduction the lump ore can well be replaced by briquettes of ore, reduction carbon and limestone. The vertical layering of the charge enables the combustion gases in the sintering zone of the ore to pass through the outer pervious coke layer. In order to restrict the sintering zone in the upward direction, top gas is blown in. The tuyeres, fed with coal dust and oxygen-enriched blast, are movable radially, in order to control the refining action. The crude steel as it flows to the rotary refining furnace is heated by a gas flame at the tap hole. In the second method, which is very similar, the entire shaft is filled with briquettes. The sintering zone is surrounded by a series of water-cooled tubes, which form an annular space for the evacuation of the gases. This form of furnace also serves to melt the ferrous oxide and slags required for the rotary refining furnace. For the smelting of acid, low-grade ores, a duplex process is proposed, in which the ore is first reduced in an acid shaft-furnace, and the crude steel is then desulphurised in a second basic shaft-furnace. From the basic furnace the metal is run into a rotary refining furnace which is also fed from the blast-furnace previously described.

The Direct Production of Steel in the Blast-Furnace. (Iron and Coal Trades Review, 1935, vol. 131, Oct. 4, pp. 545–546). An English summary of the above paper by Fränkl.

IRON INDUSTRIES.

The Saar Metallurgical Works at the Reunion. H. A. Wagner. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Aug. 3, pp. 951–958). The five iron and steel works in the Saar are described, with particular reference to their development in the years following the war. The problems of long-distance gas supply and electric current generation, as they affect these works, are discussed, and attention is given to matters affecting the technique of coking. Consideration is also given to the obtainment of vanadium from pig iron and to ore preparation experiments that have been made with a view to rendering available the dogger ores of Baden.

HISTORICAL.

Iron and Steel. Sir W. J. Larke. (Proceedings of the Royal Institution of Great Britain, 1935, vol. 28, Part 4, No. 135, pp. 690–715). The author traces the historical development of iron and steel from early antiquity to the present time. References to meteoric iron, to the smelting of iron in India as early as 1400 B.C., to the swords of Damascus and Toledo and the beginnings of an industry in Great Britain, are followed by a survey of modern development. Finally, attention is drawn to the advances which have taken place within recent years.

Man and Metal. Sir W. J. Larke. (Edward Williams Lecture, Institute of British Foundrymen, July 2 to 5, 1935: Foundry Trade Journal, 1935, vol. 53, July 11, pp. 26–29). The author traces the early history of the use of metal by man, and then discusses modern industrial development, especially the contribution of metallurgy to industrial progress.

From the History of the Production and Use of Steel. O. von Halem. (Montanistische Rundschau, 1935, vol. 27, Stahlbau-Technik, May, pp. 3–5; June, pp. 3–5; July, pp. 5–6). A short historical review of the production and use of iron and steel.

FOUNDRY PRACTICE.

GENERAL FOUNDRY PRACTICE.

Modern Foundry Practice. W. G. Morgan. (Journal of the Institution of Production Engineers, 1935, vol. 14, Sept., pp. 480–490). Modern developments in foundry technique are briefly referred to.

Design of Balanced-Blast Cupola. F. J. Cook. (Foundry, vol. 63, June, pp. 32–33, 72). The salient features of the balanced-blast cupola, in which control of the air supply to individual tuyeres is provided, are examined. It is shown that this type of cupola has many advantages over the old type, and that in many cases cupolas have been converted to balanced-blast operation with beneficial results.

Introduction of Blast through Small Tuyeres in the Shaft of Cupolas above the Melting Zone. C. Rein. (Giesserei, 1935, vol. 22, July 5, pp. 335–339). The author discusses the use of upper tuyeres in cupolas, situated above the melting zone. He comes to the conclusion that such tuyeres are of no value, and he is opposed to the opinion held by many experts, that a reduction reaction of carbon dioxide to carbon monoxide takes place in the cupola.

Cupola and Pulverised-Fuel-Fired Furnace Irons Compared. F. Boussard. (International Foundry Congress, Brussels, 1935: Foundry Trade Journal, 1935, vol. 53, Sept. 26, p. 226). The author compares the performance, under test, of cupola and pulverised-fuel-fired furnace irons.

Calculation of the Metal Charges for the Foundry Cupola. H. L. Campbell. (Metals and Alloys, 1935, vol. 6, May, pp. 107–112). The calculation of metal charges for the foundry cupola depends, in the first place, on the desired analysis of the castings to be produced. It is then necessary to determine the proportion of scrap to be used and the changes in composition due to melting; on this basis the weights of the materials to be charged may be ascertained. Special elements (such as manganese, nickel, chromium and molybdenum) may be required to be added, and any losses of the alloying metals should be allowed for. If two or more different lots of metal are available the proportions to be employed for a charge of a definite composition may be calculated by formulæ or by graphical means. The author furnishes examples of these calculations and of the use of charts in solving foundry cupola problems.

Separation of Cupola Mixtures. A. J. Shore. (Institute of British Foundrymen: Foundry Trade Journal, 1935, vol. 52, June 20, p. 413). In the author's foundry castings of a large variety of types and sizes are produced. To cover these requirements three mixtures have been chosen, which are run in two cupolas. The procedure adopted to ensure that the right mixture is used for each casting is indicated.

Metallic Losses in Cupola Melting. G. Hénon. (International Foundry Congress, Brussels, 1935: Foundry Trade Journal, 1935, vol. 53, Sept. 26, p. 235). It is suggested by the author that the metal losses in cupola practice may be assessed by calculating the weight of slag produced on the basis of the amount of limestone charged, and, by obtaining the iron content of the slag, so evaluate the metal losses. The application of this method is discussed.

Determines Iron Lost in Melting. W. McConnachie. (Foundry, 1935, vol. 63, Apr., pp. 26, 56). The author points out that the iron lost by oxidation in the cupola may be estimated by an analysis of the gas from the cupola.

Recent Developments in Foundry Melting Practice. Oxidation and Carburisation Factors. J. E. Fletcher. (Fuel Economy Review, 1935, vol. 14, pp. 27–32). A general review of the problem of oxidation in cupola working is followed by a discussion of the balanced-blast cupola.

The Soda-Ash Process in the Foundry. A. G. Lefebvre. (International Foundry Congress, Brussels, 1935: Foundry Trade Journal, 1935, vol. 53, Sept. 26, p. 234). The use of the soda-ash process for the refinement of cast iron is dealt with.

Churned Iron and Soda Treatment. L. E. C. Girardet. (International Foundry Congress, Brussels, 1935: Foundry Trade Journal, 1935, vol. 53, Sept. 26, pp. 223, 236). The method of improving cast iron by the addition of soda-ash in rotating moulds is briefly referred to.

Slags and Gases in Cupola Operation. A. H. Dierker. (American Foundrymen's Association, Aug., 1935, Reprint No. 12). The author discusses the theory of combustion in cupola operation, oxidising and reducing conditions and the factors affecting these conditions, and cupola reactions.

Sesci Rotary Melting Furnaces. (Machinery (Lond.), 1935, vol. 46, July 4, pp. 421–422). Sesci rotary furnaces consist of a barrel with cone-shaped ends, to one of which a patent burner is fitted, the other leading to the exhaust, at which end material is charged into the furnace. Pulverised coal is the fuel employed. Preheating of the air for combustion is achieved by recuperators. A neutral, reducing, or oxidising flame may be produced at will, and the melting time is quite short, a grey iron melt being poured in less than two hours.

Electric-Furnace Cast Iron—a New Engineering Material. A. E. Rhoads. (American Society for Metals, Dec. 3, 1934: Iron and Steel of Canada, 1935, vol. 18, Mar.–Apr., pp. 19–24). The present tendencies in cast-iron practice are briefly reviewed, and the types of iron now available are classified. Typical irons are described and the value of the electric furnace in the production of special grades—such as the superheated and heat-treated cast irons—is stressed.

Founding High-Test Cast Iron. W. R. Jennings. (Foundry, 1935, vol. 63, June, pp. 28–29, 66, 68). Among the factors which bear upon the production of high-tensile cast irons, the author suggests the following as being important: The quality of the materials (coke and metal); the depth of the hearth; and consistent operation. He then considers certain casting problems, such as the temperature of pouring, the design of the cores to withstand the high pouring temperatures usually employed, the avoidance of defects in castings and the correct use of gates and risers.

Heavy Alloy Iron Castings. C. C. Miller. (Metals and Alloys, 1935, vol. 6, July, pp. 185–186). Notes on the production of heavy castings in an alloy cast iron called Cecolloy, are furnished. The chief alloying element is molybdenum, to the extent of about 0·30 to 0·50 per cent. for medium castings, augmented by chromium or nickel for heavier and lighter castings.

The Production and Utilisation of Virgin Irons containing Titanium and Vanadium. J. Challansonnet. (International Foundry Congress, Brussels, 1935: Foundry Trade Journal, 1935, vol. 53, Sept. 26, p. 231). The production of cast irons containing between 0·4 and 0·7 per cent. of vanadium, and of those containing small amounts of titanium (about 0·3 per cent.) is dealt with, and the properties and uses of such irons are indicated.

Developments in the Production and Use of Refined and Alloy Pig Irons in Great Britain. G. T. Lunt. (International Foundry Congress, Brussels, 1935: Foundry Trade Journal, 1935, vol. 53, Sept. 26, pp. 222–223). This paper constitutes a survey of recent developments (the use of blending, degasification, and so on) in the production of refined and alloy pig irons.

Dust Collection in the Foundry. S. D. Moxley. (American Foundrymen's Association, Aug., 1935, Reprint No. 17). Various types of equipment available for the collection, treatment and disposal of dust in foundries are discussed and illustrated.

Protection against Floor Vibrations in Foundries. W. Gerb. (Giesserei, 1935, vol. 22, July 19, p. 364). The vibrations set up in the floor by jolting moulding machines can often cause damage to moulds. Protection is obtained, not by insulating the foundations of the machines, but by the use of insulated concrete platforms on which the moulds are stood.

MOULDS AND CORES.

Calculations of Pipe Patterns. (Röhrenindustrie, 1935, vol. 28, Sept. 11, pp. 101–102). A brief discussion of the making of wood patterns for the founding of pipes.

Prevent Losses with Proper Gates and Risers. P. Dwyer. (Foundry, 1935, Apr., vol. 63, pp. 42–44, 47). Examples of the proper application of gates and risers to the casting of malleable iron pipe fittings, chilled castings, &c., are examined. (*See* Journ. I. and S.I., 1935, No. I. p. 363).

New Compressed-Air Press Moulding Machine. Weil. (Giesserei, 1935, vol. 22, July 5, pp. 333–335). An illustrated description of the machine is given. It can be used to advantage for parts that can be moulded in boxes about 100 to 150 mm. high or for which a preliminary ramming of the sand is not required.

The Hand Ramming of Sand. F. Boussard. (International Foundry Congress, Brussels, 1935 : Foundry Trade Journal, 1935, vol. 53, Sept. 26, pp. 232–234). Among the points to which attention is drawn are the following : The relation between the degree of ramming and the mould hardness ; the work done by ramming ; the optimum ramming depth ; the shape of the rammers ; and the technique of ramming.

Automatic Mould Elevator and Conveyor Unit Offers Marked Economies. (Iron Age, 1935, vol. 136, Aug. 22, pp. 20–21). This appliance consists of a structural frame fitted with tiers for storing moulds ; each tier is composed of a series of trays 18 in. wide by 12 ft. long, each with a maximum capacity of eight 16-in. × 16-in. flasks ; a four-tier unit has a maximum capacity of 288 moulds. An elevator moves the trays of moulds to any desired tier. The moulding station is at one end of the frame and the pouring station at the other, a continuous rotation of moulds being thus provided.

The Separation of Loam Cores in Shell Form. E. Feil. (Giesserei, 1935, vol. 22, Aug. 16, pp. 401–406). After touching on the testing of loam cores, the author describes his tests made to determine the conditions under which loam cores could be separated from the casting in the form of shell-like pieces, in order that their removal should be complete and expeditious. The core mass, after completion of the casting (that is, after baking) must possess a certain degree of strength ; the strength of the facing material must be greater than the force with which the graphite blacking adheres to the surface of the casting. The adherence of the facing material to the core mass must likewise exceed that of the blacking to the casting ; among other means, this can be attained by making the strength of the mass equal to, or, better, greater than, that of the facing (both materials must have approximately the same coefficient of expansion). The

facing material must give a smooth surface ; it must therefore be fine-grained and not too fat—that is to say, after drying, it must show no cracks. The facing must not burn on to the casting ; the blacking must therefore be of good quality and adhere well to the facing material. After drying, the facing must be absorptive, to take up the blacking rapidly. The presence of carbon (organic material) in the core mass facilitates its subsequent removal in the form of shell-like pieces.

Describes Alternative Methods for Drum Cores. (Foundry, 1935, vol. 63, June, pp. 56–58). Alternative methods of making the cores for drum castings are described.

The Control of Mould-Drying Processes. V. Aronovitch. (Foundry Trade Journal, 1935, vol. 52, June 13, pp. 393–396). The arrangements so far adopted to control the drying of moulds and cores have been empirical, crude, and generally unreliable. The author gives details of a method which makes use of the change in electrical conductivity of the sand with alteration in the moisture content. Two resistance couples are inserted into the green mould or core and the conductivity throughout the drying operation is recorded. A battery is connected to a voltmeter and to the resistance couple ; when all the moisture has evaporated the conductivity becomes zero and the indicating needle of the voltmeter attached to the resistance couple returns to the zero position. The couples consist of two brass wires insulated except for the tips, which are inserted in the required position in the mould, and spaced at 5 to 10 mm. apart. By using insulated couples inserted in moulds at different depths the nature of the drying process may be revealed. The author gives an example of an experiment on those lines. The same technique has been applied to the investigation of the behaviour of a green-sand mould during pouring. It is suggested that the electrical resistance method is particularly suitable for car- and drawer-type ovens, for floor drying and for experimental work.

MOULDING SANDS.

Foundry Sands. T. R. Stanley. (Canadian Mining and Metallurgical Bulletin, 1935, Sept., No. 281, pp. 391–396). Foundry sands are discussed from the point of view of the characteristics which it is necessary or desirable that these sands should exhibit.

Method for the Valuation of Moulding Sands by the Proportion of the Grain-Sizes. Bültmann. (Giesserei, 1935, vol 22, June 21, pp. 307–308). The author describes and explains a method of graphing the results of sieve analyses of moulding sands in such a way that full information regarding the proportions of the various grain-sizes present can be obtained. The determination

of the principal grain-size and the peak value of the grain-size distribution gives values from which the quality of the moulding sand may be judged.

Advocates Round Grain Sand. D. D. Cameron. (Pittsburgh Foundrymen's Association, Mar. 18, 1935: Foundry, 1935, vol. 63, Apr., pp. 31, 62). In discussing the causes of porosity and other defects in steel castings, the author stresses the value of round grain sand; it is said to ensure a higher permeability than that of sand having angular grains, permitting harder ramming and hence better castings. He considers that the use of green sand is in many cases advantageous.

Method for the Valuation of Moulding Sands on the Basis of the Grain-Size Distribution. R. von Halasz. (Giesserei, 1935, vol. 22, Sept. 13, pp. 466–468). The author suggests that in future standards the grain-size distribution should not be represented in the manner now employed, in which the percentage of each grain-size is related to the grain-size, but that the percentage of the sand passing through the sieve should be related to the diameter of the sieve mesh. Further, he indicates a simple relation between the gas permeability and a factor representing the grain-size distribution, by means of which the permeability of a sand can be estimated from the grain-size distribution alone.

The Determination of Moisture in Moulding Sands. S. E. Dawson and F. T. Hanks. (Foundry Trade Journal, 1935, vol. 53, Aug. 15, pp. 113–114). The method adopted by the authors is based on the conductivity of the moistening medium in a similar manner to the dionic testing of water, where the percentage of conducting matter in the water may be determined by electrical measurements when suitable compensation for changes of temperature is made. The apparatus consists essentially of a special form of electrodes, used in conjunction with an ohm-meter and suitable battery, the electrodes being so designed as to establish the greatest contact with the mass to be tested. The electrodes are provided with four stems and spikes for penetrating the mass, and are cross-connected, thus causing the whole circuit to be balanced. The results obtained are presented.

New Methods of Sand Preparation. H. Herring. (Giesserei, 1935, vol. 22, Sept. 13, pp. 468–469). An illustrated description of the "Samum" sand-preparing machine is given. Old and fresh sand, and the necessary amount of coal dust, are mixed together to produce a good moulding sand in a single process.

SPECIAL CASTINGS.

Castings Moulded after the Safety Pattern. I. A. Brinkman. (Steel, 1935, vol. 96, June 17, pp. 35–38). An account is given of

some of the precautions necessary when large castings are to be produced, in order to minimise accidents.

Important Data regarding the Production of Chill-Cast Iron, with Particular Reference to the Casting of Chilled Rolls. H. Bocek. (Giesserei, 1935, vol. 22, July 19, pp. 357–364). The author first discusses the various factors to which careful attention must be paid in the production of cast iron of good quality which will chill properly: The chemical composition of the pig-iron, the structure of the chill-cast iron, the casting temperature, &c. He then gives attention to the types of furnace suitable for melting the iron, and considers various influences which may affect the strength properties and the depth of chill. It is pointed out that the charge should not contain pig irons with too widely differing silicon contents (that is, irons which will, and will not, chill), because, though the average analysis may be correct, wasters will occur. Such a charge gives rise to non-uniformity of structure (core hardening), and non-uniform chill leads to large casting stresses. The strength is much reduced, the brittleness is increased, and fissuring is promoted.

Notes on Roll Failures with Reference to Chilled Cast Iron Rolls Containing Chromium and Molybdenum. J. S. Caswell. (Proceedings of the Institution of Mechanical Engineers, 1934, vol. 128, pp. 459–467). Details are given of some fractures of chromium-molybdenum iron rolls, and these are compared with the fractures usually associated with ordinary chilled iron rolls. The probable causes of fracture of the chromium-molybdenum rolls are discussed in the light of the quality and structure of the iron, the design of the rolls, and the effect of thermal stress. It is suggested that the causes of failure were (1) low strength of the iron, resulting from excess of chromium, (2) the defective design of the roll wobbler, and (3) the thermal stresses set up in the rolls during service. Detailed consideration is given to the effect of the wobbler design in preventing the free contraction of the roll after casting. It is suggested that the interference set up produces definite planes of weakness in the rolls. Particulars of an unusually long life of an ordinary chilled iron roll are given, together with a comprehensive analysis of the iron in this roll.

Roll Production. (Machinery (Lond.), 1935, vol. 46, May 30, pp. 253–259). In this article the practice of Sir W. G. Armstrong Whitworth and Co. (Ironfounders), Ltd., Close Works, Gateshead-on-Tyne, in the founding of rolls is described. The melting plant consists of two air furnaces fired by pulverised fuel, and four Sesci furnaces. Both chilled and grain rolls are made; the former are of the carbon, molybdenum, nickel, or nickel-chromium alloy type, and are marketed under the name of Closeloy. Sizes up to 42 in. in diam. and 108 in. long, weighing up to 25 tons may be produced. The main machine shop equipment comprises four

large- and medium-sized roll-turning lathes, a wobbler end milling machine and a roll grinding machine.

Melts Iron in Air Furnaces. P. Dwyer. (Foundry, 1935, vol. 63, July, pp. 18–20, 44, 47). An account is given of the practice at the plant of the Mackintosh-Hemphill Co., Pittsburgh, for the production of sand-cast and chilled-iron rolls. The majority of the rolls are cast from iron melted in air furnaces. The mixtures used are tabulated.

Hardens Steel Roll Surfaces. P. Dwyer. (Foundry, 1935, vol. 63, Aug., pp. 36–37, 77–78). The practice at the plant of the Mackintosh-Hemphill Co., for the production of carbon and alloy steel rolls is illustrated and described. A process has been developed for imparting a hard surface to the face of the rolls by means of a torch and spray.

The Manufacture of Piston Rings. J. Challansonnet. (Foundry Trade Journal, 1935, vol. 53, Oct. 17, pp. 295–297). After discussing wear and lubrication in general terms, the author proceeds to examine the requirements of piston rings and the methods of casting them. The two methods of casting commonly employed —as cylinders or as single rings—are examined, and suitable chemical analyses for the irons are suggested. Attention is also given to the defects from which piston rings suffer, and the methods of introducing the " spring " into these rings.

A New Multi-Casting Process. R. Leonhardt. (Foundry Trade Journal, 1935, vol. 52, June 13, pp. 397–398). The multi-casting process, while possessing many advantages, suffers from certain disadvantages—for instance, the fact that comparatively shallow articles only can be produced, and that trouble with the runners is sometimes encountered. A new process has been devised and patented by Badische Maschinenfabrik, of Durlach, in which the flasks are superimposed in the usual way, but in which the runners are arranged alternately to the right and left of the assembly, thus enabling the flasks to be cast separately, while lifting due to high ferrostatic pressure is eliminated. A new type of squeeze machine has been designed for the moulding operation, actuated by compressed air ; this machine is described.

Pages from a Foundryman's Notebook. II.—Making Drumsides in Loam. (Foundry Trade Journal, 1935, vol 53, July 4, p. 14). In this article the technique of moulding the sides of a drum for a hoisting engine is described.

Moulding a Rope-Driven Pulley Wheel. F. Oldershaw. (Iron and Steel Industry, 1935, vol. 8, Aug., pp. 422–425). The procedure to be adopted in moulding a grooved pulley wheel is explained.

Production of a Cylinder Cover without the Use of Chaplets. F. Vogelsang. (Giesserei, 1935, vol. 22, Aug. 2, p. 387).

Strickle Moulding of Large Castings. H. Fabre and H. Dubois. (Institute of British Foundrymen, July 2 to 5, 1935: Foundry Trade Journal, 1935, vol. 53, July 11, pp. 34–36; July 18, pp. 51–54). A number of applications of strickle moulding to large castings is cited. These include a blackpot weighing 12 tons, a rolling-mill roll, a large pipe bend, and a four-bladed propeller; the procedure for making the patterns and the moulds is indicated in each instance.

Cast Iron Plate Roofing—A New Product of the Foundry. (Iron Age, 1935, vol. 136, Sept. 12, pp. 10–14). The roofing is manufactured by the United States Pipe and Foundry Co., Burlington, New Jersey, U.S.A., and marketed under the name of Usicast. It is composed essentially of two main units—a standard cast iron plate and a cap plate. The former is a single casting $\frac{5}{16}$ in. thick, 52 in. long and 24 in. wide, weighing 78 lb. The cap plate is semi-cylindrical in shape and of the same length as the roof plate. Usicast roofs are stated to be flexible in erection and permanent in use.

Prevent Porosity in Fly-Wheel Hubs. H. H. Eastham. (Foundry, 1935, vol. 63, May, p. 31). It is sometimes difficult to avoid porosity in the hubs of fly-wheel castings. By using a thin sheet steel casing round the sand core a slight chilling effect was imparted to the inside of the hub, and this was effective in preventing porosity.

CENTRIFUGAL CASTING.

Production of Centrifugal Grey Iron Castings in Water-Cooled Moulds. H. W. Stuart. (American Foundrymen's Association, Aug., 1935, Preprint No. 5). The author describes the De Lavaud process for the manufacture of cast-iron pipe, and explains in detail the various operations and equipment. He then describes the apparatus used to coat the steel moulds with a pulverulent material to prevent chilling of the pipe. A comparison is also made of the properties of cast-iron pipe produced with and without the mould coating.

Heat Treating Cast-Iron Pipe. J. B. Nealey. (Heat Treating and Forging, 1935, vol. 21, Aug., pp. 382–384). **Pipe Cast Centrifugally Without Chill.** J. B. Nealey. (Iron Age, 1935, vol. 135, May 23, pp. 14–16, 94–96: Industrial Heating, 1935, vol. 2, Aug., pp. 411–415). In the De Lavaud system of pipe casting, which is described, a water-cooled rotating steel mould is employed, into which the stream of molten cast iron is poured by means of a water-cooled trough. The mould is gradually moved in an axial direction whilst rotating, and thus receives a uniform coating of

iron. The issuing stream of iron is preceded by a thin stream of ferro-alloy powder; this material coats the mould in advance of the molten metal and so inhibits the chilling effect which would otherwise take place. On removal from the mould the pipes are subjected to an annealing treatment in a conveyor-type gas-fired furnace.

Newest Plant for Centrifugal Casting of 30 in. Pipe. J. B. Nealey. (Industrial Heating, 1935, vol. 2, May, pp. 247–250, 263–264). A description of the "Mono-Cast" method of centrifugal casting adopted by the American Cast Iron Pipe Co., is furnished. In this system the pipe is cast in a sand-lined or refractory mould without a core (green-sand moulds are in fact used). Pipes from 14 to 30 in. in diam. and up to 16 ft. long are cast by this method.

Slush Pump Piston Cores Produced by Centrifugal Casting Process. A. E. Falk. (American Foundrymen's Association, Aug., 1935, Preprint No. 7). The author discusses the various factors and production problems involved in the centrifugal casting of the piston cores used in slush pumps.

STEEL FOUNDRY PRACTICE.

Suggestions for Research on Steel Castings. J. H. Andrew. (Institute of British Foundrymen, July 2 to 5, 1935: Foundry Trade Journal, 1935, vol. 53, July 18, p. 49). The author suggests that among the subjects to which research might well be directed are the following: Methods of feeding castings, the reduction of fluid-solid contraction, the use of oxygen-free steel, the types of facing sand, and the fluidity of castings.

Steel Castings, with Special Reference to Alloy Steels. W. H. Hatfield. (Institute of British Foundrymen, July 2 to 5, 1935: Foundry Trade Journal, 1935, vol. 53, July 18, pp. 43–46; July 25, pp. 69–71). The author discusses low- and high-alloy steel castings, and particularly the range of mechanical properties which can be produced by the use of suitable alloying elements. He lists the analyses, heat treatments, mechanical properties, and applications of a number of steels, including heat-, acid-, corrosion- and wear-resisting alloys. Physical data, such as the contraction allowance, freezing range and distinctive properties, are also set forth. The author goes on to consider the characteristics of these alloys from the foundry point of view, and cites some of the features—including coefficient of expansion, thermal conductivity, tenacity at high temperatures—which are of importance in the production of sound castings.

A Note on the Influence of Temperature Gradients in the Production of Steel Castings. G. A. Batty. (Institute of British

Foundrymen, July 2 to 5, 1935 : Foundry Trade Journal, 1935, vol. 53, July 25, pp. 65–67 ; Aug. 1, pp. 88–89 : American Foundrymen's Association, Aug., 1935, Preprint No. 10). It is shown that under certain conditions an adverse temperature gradient—the top of the casting being coolest and the bottom hottest—may occur in pouring difficult castings. The influence of the methods of venting and gating is examined, with reference to typical examples ; methods of pouring, including the partial and total reversal systems, are also considered. The temperature gradients which occur under conditions of partial or total reversal are suggested, and the control of solidification by this means and by the use of chills receives brief attention.

The Utilisation of Electric Arc Furnaces. M. Boutigny. (International Foundry Congress, Brussels, 1935 : Foundry Trade Journal, 1935, vol. 53, Sept. 26, p. 234). The use of the electric arc furnace in the production of steel for castings is considered.

Cast Steel Crankshafts. (Automobile Engineer, 1935, vol. 25, July, pp. 269–270). The procedure adopted at the foundry of the Ford Motor Co. of Canada Ltd., at East Windsor, Ontario, for the production of cast steel automobile engine crankshafts is described. Metal for the castings is prepared in a five-ton electric furnace, the charge consisting of 40 per cent. steel scrap, the balance being pig iron, back stock and alloys. The conveyor system is in use, permitting an output of 475 crankshafts per day, in five heats. After being cleaned the cast crankshafts are submitted to a double heat treatment in a continuous furnace. Before being machined a drop test is applied, and after rough and finish machining torsion tests are carried out. The metal used has the following analysis : Carbon, 1·35–1·60 ; manganese, 0·50–0·60 ; silicon, 0·85–1·10 ; chromium, 0·40–0·50 ; copper, 1·50–2·0 ; phosphorus, 0·10 max. ; sulphur 0·06 max. per cent. Owing to the fact that cored crankpin journals are used, the cast crankshaft is 10 lb. lighter than the corresponding forged article ; a saving in machining operations and an improvement in physical properties are also realised.

From the First Bell of Cast Steel to the Olympia Bell. H. Kalpers. (Giesserei, 1935, vol. 22, Sept. 13, pp. 469–473). The Olympic Games to be held in Berlin in 1936 will be " rung in " by a cast-steel bell. The author outlines the difficulties that have stood in the way of the recognition of steel as a suitable material for bell-founding, and then describes briefly the moulds used and the testing of the tone of the bells. He states that the first steel bell was cast by Jakob Mayer, a director of the Bochumer Verein, and was shown at the Paris Exhibition of 1855.

Small Acid Open-Hearth Furnace has Capacity of One-Half Ton. (Foundry, 1935, vol. 63, May, pp. 22–60). **Small Acid Open-Hearth Furnace has One-Half Ton Capacity.** J. Trantin,

jun. (Steel, 1935, vol. 97, pp. 50–52). Details of a small acid open-hearth furnace for experimental work are furnished. The regenerators are disposed above the furnace, giving a very short travel from the ports to the checkers. Air is supplied to the regenerators under pressure, and it is possible to operate the furnace at, above, or below atmospheric pressure. The furnace is oil-fired with spent and reconditioned lubricating oil. Heats of from 100 to 1,000 lb. may be melted, the time required for a 300-lb. heat (from charge to tap) being about 45 min.

MALLEABLE CAST IRON.

Malleable Cast Iron. D. P. Forbes. (Foundry, 1935, vol. 63, May, pp. 23, 63). The method of production, mechanical properties and industrial applications of malleable iron are briefly dealt with.

Malleable Cast Iron. J. V. Murray. (Proceedings of the Staffordshire Iron and Steel Institute, 1933–1934, vol. 49, pp. 73–79). The author considers the composition of blackheart and whiteheart malleable and its relation to the mechanical and other properties, and goes on to discuss the use of the rotary furnace in the production of malleable iron, the annealing of malleable castings, the defects caused by faulty annealing and the changes which take place. He concludes with the results of some tests on whiteheart malleable cast iron subjected to various heat treatments.

Notes on the Production of Cupola Malleable Castings. F. B. Riggan. (American Foundrymen's Association, Aug., 1935, Preprint No. 19). The author discusses the advantages and disadvantages of cupola melting, and describes the raw materials used, cupola charging and melting control.

High-Speed Fettling. (Machinery (Lond.), 1935, vol. 46, June 6, pp. 293–294). The fettling plant at the works of Ley's Malleable Castings Co., Ltd., is described. This concern produces blackheart malleable castings, mainly for the automobile trade. Large batteries of double-ended grinding machines are installed, the grinding wheels being made of Aloxite Redmanol.

FOUNDRY PLANT.

The Works of the Sheepbridge Stokes Centrifugal Castings Company, Limited. (Foundry Trade Journal, 1935, vol. 52, June 27, pp. 429–430). The plant deals with centrifugal castings for motor-car engines, piston ring drums, &c., and high-duty

alloy cast irons suitable for nitrogen hardening and heat and corrosion resistance. Four mechanically-charged cupolas are in use, delivering metal to a large oil-fired receiver of 2 tons capacity; this arrangement enables uniformity in composition to be maintained. Comprehensive centrifugal casting equipment is also installed.

Newton Chambers and Company, Limited. (Foundry Trade Journal, 1935, vol. 52, June 27, pp. 430–431). A short description of the iron founding department at the works of the above company is furnished.

New Foundry at the Rochester Works of Messrs. Winget, Ltd. (Engineering, 1935, vol. 140, July 12, pp. 35–38). **A Modern Foundry.** (Engineer, 1935, vol. 160, July 12, pp. 44–45). **The New Foundry of Messrs. Winget, Ltd.** V. C. Faulkner. (Foundry Trade Journal, 1935, vol. 53, Oct. 24, pp. 305–308.) The plant described is designed for mass-production work, the material used being known as Meehanite. This metal is a cast iron with a calcium silicide admixture, and it is said to produce a grey iron having excellent physical properties, enabling it to be adopted for parts such as motor-car engine crankshafts, camshafts, and press dies. The three cupolas employed are stationed close to the mould conveyor, and this in turn is served by a row of moulding machines. A sand-conditioning plant is linked to the storage hoppers, feed hoppers and disintegrator by conveyor belts. The core-making and fettling departments have not yet been fully mechanised, but this will eventually be done. The foundry is said to be well suited to the rapid production of light castings in any of the five grades of Meehanite which are made.

A Modern High-Output Steel Foundry. (Foundry Trade Journal, 1935, vol. 53, Sept. 12, pp. 195–198). The layout and equipment of the steel foundry of Kryn and Lahy, Ltd., Letchworth, are described and illustrated.

Modern Foundry Practice. (Machinery (Lond.), 1935, vol. 46, Sept. 5, pp. 685–689). An illustrated account is given of the equipment and methods employed at the foundry of John Harper and Co., Ltd., Willenhall.

Modern Equipment Features New Ford Foundry in Canada. (Foundry, 1935, vol. 63, Aug., pp. 42–45, 86, 89). The equipment and practice at the new foundry of the Ford Motor Co. at East Windsor, Canada, are described and illustrated.

Melt Metals with Natural Gas. P. Dwyer. (Foundry, 1935, vol. 63, May, pp. 24–26, 64). The foundry plant of the Homestead Valve Co. at Homestead, Pa., U.S.A., is briefly described. The plant is devoted to the manufacture of all types of valves; steel castings are obtained from outside sources, but the non-ferrous and grey-iron castings are melted and cast in the foundry. The

salient feature of the latter is the melting equipment, consisting of a tilting open-flame furnace for melting iron and one tilting and one stationary furnace for non-ferrous work. All the furnaces are fired with natural gas.

Production of High-Grade Castings for Machine Tools. D. M. Gurney. (Steel, 1935, vol. 97, Sept. 9, pp. 32–35). Some notes on the foundry plant of the Sterling Foundry Co., Ohio, U.S.A., are furnished. The plant produces high-grade grey-iron castings for machine tools.

COOLING PHENOMENA AND DEFECTS IN CASTINGS.

Surface Examinations in the Foundry. G. Stade. (Giesserei, 1935, vol. 22, July 5, pp. 339–340). **A New Method of Surface Examination in the Service of Corrosion Protection.** A. Kufferath. (Korrosion und Metallschutz, 1935, vol. 11, June, pp. 133–136). A new form of comparison microscope is illustrated and described.

Sound and Unsound Steel Castings. J. Obrebski. (International Foundry Congress, Brussels, 1935: Foundry Trade Journal, 1935, vol. 53, Sept. 26, p. 226). The author refers more particularly to intercrystalline cracking, which defect he discusses.

Studies on Solidification and Contraction in Steel Castings. II.—Free and Hindered Contraction of Cast Carbon Steel. C. W. Briggs and R. A. Gezelius. (Transactions of the American Foundrymen's Association, 1934, vol. 42, pp. 449–476).

Spongy Iron and the Léonard Effect. M. Kagan. (International Foundry Congress, Brussels, 1935: Foundry Trade Journal, 1935, vol. 53, Sept. 26, pp. 224–226). The author discusses the nature and causes of sponginess in iron, and of the Léonard effect (sponginess due to local and excessive heating of protruding portions of sand).

PRODUCTION OF STEEL.

STEEL-MAKING PRACTICE.

Modern Crucible Melting Equipment. R. H. Stone. (Industrial Heating, 1935, vol. 2, Apr., pp. 181–186). The subject is considered under the following headings: (1) The furnaces, (2) the fuels and burners, (3) the crucibles, (4) the metal produced, (5) the costs. It is shown that modern crucible furnaces are often heated with fuel oil or gas, in which case tilting furnaces may be used. Special refractories of the silicon carbide type are employed for the furnace lining, but graphite continues to be favoured for the crucibles. The author also compares the costs for coke-, oil- and gas-fired furnaces.

Manufacture and Properties of Bessemer Steel. C. C. Henning. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 623). After discussing the present status of Bessemer steel production in America, the author enumerates the uses of Bessemer steel, giving typical analyses for different applications, and describes the outstanding characteristics of the Bessemer process. The properties of the material are also described.

The Developments in the Design and Operation of Open-Hearth Furnaces. R. S. Bower. (Iron and Steel Engineer, 1935, vol. 12, July, pp. 413–423). The author traces the history of the open-hearth furnace, giving an account of the development of the furnace itself (in point of size), the regenerators and the charging mechanism. Consideration is then given to meters, reversing valves and furnace insulation, and also furnace controls. Finally, an indication is given of the possible future trends in furnace design.

Improving Open-Hearth Furnace Design. W. C. Buell, jun. (Steel, 1935, vol. 96, June 3, pp. 56–58; June 10, pp. 52–56). The series of articles under the above title treats of the equipment (regenerators, waste-heat boilers, flues, &c.), which are complementary to the open-hearth furnace proper. In the first article the author describes the layout of the open-hearth furnace below the charging floor line, and gives the disposition of the various items—regenerators, slag pockets, gas flues and so forth. The principles which should be followed in designing the below-furnace layout for a 150-ton furnace are expounded. In the second article essential data and diagrams for the layout and

operation of the slag pockets serving the regenerators are detailed. The entrained matter which is deposited in the slag chambers may be of two kinds: Heavy material and light particles. Excessive carry-over of solid matter is, of course, undesirable. The method of dealing with the accumulated slag is indicated.

Improving Open-Hearth Furnace Design. W. C. Buell, jun. (Steel, 1935, vol. 96, June 17, pp. 49–52; June 24, pp. 50–54; vol. 97, July 1, pp. 40–46; July 8, pp. 50–53; July 15, pp. 57–61; July 22, pp. 45–48). In an article dealing with slag pockets, data are provided for the calculation of the slag pocket size, and the effect of improvements in design is indicated. Attention is then devoted to the refractories used in slag-pocket construction, and methods of effecting economies in this direction are suggested. The actual design of the slag pockets is then dealt with, particularly the construction of the arches to the pockets. In a subsequent article the detailed design of the steelwork for the slag pockets is treated, together with some special constructions which have been tried and found wanting. The subject of fantails (connecting flues) then receives attention. Fantails are used to connect the slag pockets and regenerator chambers. The layout and constructional details of typical fantails are described, and the various factors which influence design are examined.

Improving Open-Hearth Furnace Design. W. C. Buell, jun. (Steel, 1935, vol. 97, July 29, pp. 40–45; Aug. 5, pp. 58–62; Aug. 12, pp. 54–56). In the first article on regenerators the dimensions of typical regenerator chambers (exclusive of the checkerwork contained therein) are given. The subsequent instalment treats of regenerator walls and roofs; data relating to various types of construction are presented. In the third article on this subject the various details of construction are examined, and the superiority of insulated over uninsulated chambers is demonstrated.

Improving Open-Hearth Furnace Design. W. C. Buell, jun. (Steel, 1935, vol. 97, Aug. 19, pp. 44–52; Aug. 26, pp. 45–50; Sept. 2, pp. 63–67). The detailed design of the flues connecting the regenerator outlets with the flue system proper is analysed. Two articles on the design of regenerator chambers are next presented, in which detailed information on the construction of the brickwork and steelwork for these chambers is set forth.

Relationship between the Type of Port, the Output and the Oxidising Effect in Open-Hearth Furnaces. F. Wesemann. (Stahl und Eisen, 1935, vol. 55, Sept. 12, pp. 981–989; Sept. 19, pp. 1006–1009). On the basis of operating experience with various open-hearth furnaces and fuel-technological considerations the author sets out the best dimensions for the ports of such furnaces. The resulting course of the combustion of the gas influences not only the output but also the oxidising effect of

the furnace—the term " oxidising effect " (*Frischwirkung*) is used in the more restricted sense of the scaling of the charge by the flame gases. It is to be assumed that between this oxidation of the scrap and the change in the rate of decarburisation and of the ferrous oxide content of the bath there exist close relationships, which require further elucidation. An experiment, intended to show the influence of the excess of oxygen in the flame on the output and oxidising effect, demonstrated that a reduction of the amount of excess air impairs the furnace output—hence the considerable influence of factors depending on the mode of flow of the gases (for instance, the position of the flame) on the output. Finally, it appeared, from the replies to an enquiry addressed to a number of open-hearth steelworks, that a close connection does actually exist between the dimensions of the port openings and the scaling of the charge, and its nature is discussed. It is apparent that the relations between the design of the port, the mode of combustion, the output, the oxidation and the reactions occurring during the boil need further investigation.

Operating Results with Open-Hearth Furnaces of the Terni Type. O. Göbel. (Stahl und Eisen, 1935, vol. 55, Aug. 15, pp. 882–889). After touching briefly on the principles of the Terni open-hearth port, the author gives details of the results obtained at seven works where such furnaces are installed (the Terni Works and two others in Italy, two in France, one in England and one in Poland). Comparison of the results shows that the variations in the outputs, heat consumptions and lives of the furnaces are due to local operating conditions. A method of cooling the port by means of cooled clean waste gases, by which the life of the furnace is considerably increased, is described. The author adds that these figures cannot be compared with those for German furnaces, because account cannot be taken of the effect of the efficiency of the furnace crew on the results obtained.

Gaseous Fuels in Modern Open-Hearth Furnace Practice. J. H. Hruska. (Industrial Heating, 1935, vol. 2, Feb., pp. 87–90 ; Mar., pp. 143–144 ; Apr., pp. 199–200). Producer gas is the fuel principally dealt with. The characteristics of this fuel are first considered, and this is followed by an analysis of the heat balance in open-hearth furnaces ; finally, the advantages of hot and cold gas are compared.

The Raw Materials for the Open-Hearth Steelworks and their Use according to the Lessons of the War and Post-War Periods. L. Granger. (Revue de l'Industrie Minérale, Mémoires, 1935, No. 351, Aug. 1, pp. 365–371). The author discusses the influence of the quality of the raw material charged into the open-hearth on that of the steel produced, basing his remarks on his own experience during and after the war. During the war steel turnings were used for making the synthetic pig, manganese pig

and spiegel for the manufacture of steel with complete success, but after the war the quality of the steel produced depreciated. The author attributes this to the use of oxidised (badly rusted) turnings and scrap. In passing, he mentions that the pig iron made from such scrap was also useless for puddling.

Investigation on Steelworks Tar. H. J. van Royen, H. Grewe, and K. Quandel. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, May, pp. 479–488). The authors have made tests to check the accuracy of various methods used for determining the principal characteristics of steelworks tar. Of those examined, only the methods for estimating the specific gravity, the softening point of the pitch and the naphthalene content were sufficiently correct. The others—the fractional distillation test, the indentation test and the determination of the free carbon—gave very scattered results when carried out in different laboratories. Accordingly, the sources of error in these tests were sought out, and improved methods were worked out.

Residual Metals in Open-Hearth Steel. J. D. Sullivan. (Metals and Alloys, 1935, vol. 6, May, pp. 134–136). Owing to the use of alloy steel scrap, tinplate and galvanised scrap in open-hearth charges, the copper, chromium, nickel and tin contents of the steel produced in America have been gradually increasing. Samples of steel produced at a number of American plants have been examined at the Battelle Memorial Institute and particulars are presented showing the amounts of residual metals present in the samples.

Contribution on the Optical Measurement of the Temperature of Molten Iron and Steel. F. Blaurock. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, June, pp. 517–532). Of the two chief methods of optical pyrometry—total and partial radiation pyrometry—only the latter is of interest in the measurement of the temperature of molten iron and steel, particularly in " open " furnaces ; colour pyrometry, which in theory also offers useful possibilities, is still awaiting the development of more perfect apparatus. The disadvantage of the optical method is that frequently the emissivity of the spot sighted on is not known precisely, and consequently the temperature correction is also uncertain. From a survey of the literature and the results of experiments, it would appear that errors in partial radiation pyrometry can be avoided, or at least minimised, in three ways : (1) By the use of a coloured filter of a wave-length shorter than that of red light ; (2) by measurement in two colours consecutively (with a filament-type pyrometer) ; and (3) by working out more precise details for the measurement in red light. Measurement in the green confirmed the conclusions to be drawn from the literature. The relative sensitiveness of the eye is, in practice, greater to green than to red light, and, further, the emissivity

appears to increase as the wave-length decreases—evenly if only slightly; thus, the temperature correction for green light is about 20 per cent. less than for red light. In the two-colour method, measurements are made in the red and then in the green; from the difference between the values obtained, the correction can be determined. The accuracy of the measurements is about $\pm$ 15°, but very exactly standardised instruments and some practice are necessary.

Previous experience of optical measurements has been extended and summarised. It is not possible to establish two fixed values for the emissivity, one corresponding to oxide-free iron and the other to oxidised iron, because the values for the surface can vary between 0·4 and unity; the best that can be done is to grade into two groups corresponding to the two conditions mentioned, namely, 0·4 to 0·68 (dark) and 0·68 to 1 (light). Even at precisely defined positions of measurement and under apparently similar conditions, the emissivity may vary within the limiting values of these two groups; in general, however, these variations are less wide and lie close to certain values in practice, which can then be made the basis for calculation. The question whether "dark" or "bright" spots should be sighted on can be decided only at the particular measuring point itself; in this connection it must be remembered that the eye can only compare brightnesses, so that "bright" and "dark" are not absolute values, and also "dark" always corresponds to the oxide-free condition, whereas "bright" may be due to oxidation, to reflection or to both. In observations on the stability of the oxide or slag coating on cast iron at temperatures of 1,360° to 1,500° C., it appeared that the coating disappeared at higher temperatures; the temperature of this change seemed to depend in some degree on the manganese and silicon contents. Exact numerical data on this matter are not available, owing to the possible variations of composition and differences between the measuring points in making the determination in practice. The composition had an influence in helping or hindering the formation of the oxide coating, but no clear connection with the emissivity could be found.

In the third part of the paper the author has collected together all the data and experience available in the form of instruction cards, in which are included all the information necessary for the carrying-out of partial radiation pyrometry in the iron and steel works.

Equilibrium in the Reaction of Hydrogen with Ferrous Oxide in Liquid Iron at 1,600° C. M. G. Fontana and J. Chipman. (American Society for Metals, Oct., 1935, Preprint No. 29). An experimental study of the reaction of hydrogen with dissolved ferrous oxide in liquid iron has led to the discovery and elimination of a source of error which had affected the accuracy of the

results previously reported. It was found that the equilibrium constant of the reaction is a true constant at all oxygen concentrations up to saturation. The results are used to obtain the free energy of ferrous oxide at 1,600° C. Calculations show that the results of this investigation constitute a confirmation of Vacher's experiments on the iron-carbon-oxygen system.

The Equilibria between Iron and Nickel and their Silica-Saturated Silicates. P. Bardenheuer and E. Brauns. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1935, vol. 17, No. 9, pp. 127–132). By means of melting tests in sand crucibles the equilibria between iron-nickel alloys and iron-nickel silicates at a temperature of about 1,600° C. have been determined. The distribution of the nickel and iron between the metal and the silicate slag followed approximately the ideal law of mass action, although the reaction concerned was not a simple double decomposition between the lower-oxide silicates, for noticeable amounts of ferrous oxide occurred in the slag. Particular attention was paid to the solubility relations between the metals and slags. It was found that the oxide content was determined almost solely by the ferrous oxide in the case of iron-rich melts and by the nickel oxide in the case of nickel-rich alloys. Nickel contents of up to 50 per cent. were almost without influence on the solubility of ferrous oxide in iron-rich melts, and the partition coefficient [O]/(NiO) for nickel-rich melts was hardly affected by iron contents of up to 5 per cent. The solubility of nickel oxide in nickel-rich alloys diminished with decreasing temperature.

Investigations on the Course of the Metallurgical Reactions in the Basic Open-Hearth Process. P. Bardenheuer and G. Thanheiser. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1935, vol. 17, No. 10, pp. 133–147). In order to investigate the course of the metallurgical reactions taking place in the basic open-hearth process, the changes of concentration, particularly of oxygen, in the metal and slag were observed in eight heats. Halfway through the heat the reaction between the carbon and oxygen takes place ; at higher carbon contents and temperatures it proceeds very rapidly. So long as sufficient carbon (not less than 0·2 per cent.) is present in the steel and can react, it alone is practically the sole factor determining the oxygen content of the bath ; as it sinks below this value, the oxygen content of the steel becomes more and more controlled by the ferrous-oxide content of the slag. It follows that, for the production of low-carbon steel, a finishing slag low in ferrous oxide is essential ; in other words, the slag must be so worked that its ferrous-oxide content, which may be high to start with to suit the carbon content of the bath, will decrease with the passage of time. The carbon-oxygen reaction is hindered by too low a concentration of one or both components, by too low a temperature

and by excessive viscosity of the slag; incorrect measures to speed it up may easily lead to an oxygen-rich end product. In order that the metal may be degassed the boil must be sufficiently long and strong. The contents of manganese in the bath and of manganous oxide in the slag control the ferrous-oxide content of the slag and consequently the oxygen "offered" to the bath and the rate of boiling; a high manganese content hinders the excessive rise of the oxygen content in a low-carbon bath. The return of manganese from the slag is a sure sign of a low oxygen content in the bath; the greater the amount of manganese reduced out of the slag by a low-oxygen bath, the less becomes the need for subsequent deoxidation. The addition of ferro-manganese or other ferro-alloys to deoxidise the bath entails the great danger that dissolved gases are also introduced into the metal, and at a moment when they can no longer be removed by the boil; the total content of oxides can be increased instead of decreased in this way. Manganous oxide in finished steel can rise only very slowly; on teeming, it attacks the siliceous materials in the ladle lining, &c., and gives rise to silicate or sand inclusions.

The Distribution of Phosphorus between Metal and Slag in the Basic Process of Steel Manufacture. E. Maurer and W. Bischof. (Iron and Steel Institute, Sept., 1935; this Journal p. 13).

The Action of Carbon as a Reducing Agent in the Reactions of the Steelmaking Processes with Acid Slags. F. Körber and W. Oelsen. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1935, vol. 17, No. 4, pp. 39–61). On the basis of the results of a series of experimental melts made in sand crucibles, the directions in which the reactions between low-carbon iron melts, molten ferrous-oxide/manganese-oxide/silicates and solid silica proceed are first discussed. The reduction of silicon by iron and manganese from fluid silicates and from the solid silica proceeds relatively rapidly and without hindrance nearly to equilibrium, while the oxidation of silicon by thin ferrous-oxide-rich silicate may be almost completely suppressed by the formation of a solid skin of silica between the iron melt and the slag. On the basis of the experiments on the equilibria between low-carbon melts, acid slag and solid silica together with the equilibrium conditions requisite for the decarburisation reaction, the requirements for reaction equilibrium in the presence of carbon in steel melts for temperatures of 1,500° and 1,600° C., are worked out and discussed. A large number of tests in sand crucibles show that deviation of the compositions of the steel melt and slag from the equilibrium concentrations of the decarburisation reaction is a necessary but not a sufficient prerequisite for it to set in; the conditions for the setting-in and the delay of decarburisation in melts in small sand crucibles under a slag are indicated; the character of this reaction is discussed, and it is shown

that, because it results in the formation of a gas phase from a liquid phase, its course is very much influenced by external conditions, particularly those which affect the formation of the first carbon monoxide bubbles. With regard to the simultaneous procedure of the reactions of iron, manganese, silicon and carbon with the oxides in the slag, the lining and the molten steel, the elements iron, manganese and silicon can attain their equilibrium concentrations in the melt and in the slag only if the carbon reaction proceeds very slowly. If the reaction of the carbon is lively, then equilibria of the iron, manganese and silicon with their oxides in the slag, lining and melt are disturbed ; the carbon then acts as such a strong reducing agent that all the other reactions become subordinate. In particular, if the carbon reacts vigorously, the oxidising action of silicate slag on the steel melt may be to a great extent suppressed ; the carbon may then react with the solid silica of the crucible—the " crucible reaction." The course of this crucible reaction for low-manganese steel melts with various carbon contents and for melts in which both the carbon and the manganese contents vary, is outlined and discussed.

Recent Progress in Steel Making Reported from Germany. S. Epstein. (Metals and Alloys, 1935, vol. 6, May, pp. 113–117 ; June, pp. 160–164 ; July, pp. 181–184). An English summary of the work of various German investigators dealing with slag-metal equilibria, gases in steel, desulphurisation and other problems in steel making. With regard to hydrogen it is suggested that the gas probably plays a part in the rimming action, and may accumulate in the interstices between the dendrites. Its presence may give rise to internal stresses, but since hydrogen diffuses easily, annealing should greatly reduce its harmful effect. A vigorous boil helps to eliminate hydrogen from the steel bath. With respect to carbon monoxide, to which the rimming action is chiefly due, manganese appears to play an important part in controlling the gas evolution, and hence the rimming action. Nitrogen owes its interest mainly to the ageing effect to which it gives rise. The influence of the furnace gases on the sulphur content of the steel is discussed, with particular reference to the reactions between the sulphur, the iron and the oxygen ; in this connection the work of Maurer and Bischof is quoted. The slag reactions—more especially the equilibria between CaO, SiO_2, and sulphur—are also considered.

The Deoxidation of Open-Hearth Steel. G. Husson. (Revue de l'Industrie Minérale, Mémoires, 1935, No. 349, July 1, pp. 317–324). This article is divided into three parts. In the first, the author discusses the mechanism of the action of deoxidisers when added to molten steel ; in the second, he deals with oxide inclusions ; and in the third, he gives consideration to the relation between the degree of oxidation of the steel and that of the slag.

The Desulphurisation of Pig Iron and the General Laws Governing the Desulphurisation of Iron. H. Wentrup. (Iron and Steel Institute, Carnegie Scholarship Memoirs, 1935, vol. 24, pp. 103–166). Insufficient scientific data have hitherto been available to allow of the desulphurisation of pig iron to be described in quantitative terms. It is the aim of the present work to lay down the required principles, starting from practical observations of the desulphurisation of pig iron that takes place in the ladle and mixer, and proceeding to a statement of the laws governing the desulphurisation of iron in general. Attention is paid to the desulphurisation of iron by solution of the sulphur in a slag, by a diminution of the power of the iron itself to dissolve sulphur, by the formation within the iron of insoluble sulphides, and by combinations of these separate methods of desulphurisation. In particular, an explanation is given of the part played by carbon, silicon and phosphorus associated with manganese in the iron, and of the effect of the temperature on desulphurisation as learned from recent experiments. It emerges that to ensure desulphurisation these elements must be present, and the iron must be cooled.

These principles having been established, it becomes possible to throw a clearer light on the processes attending the desulphurisation of pig iron and on the possibilities of controlling them.

Sulphur in Steels and Cast Irons. H. Thyssen and W. Bonhomme. (Revue Universelle des Mines, 1935, vol. 11, July, pp. 316–325). The features of the Fe–C–S, the Fe–FeS–Mn–MnS and the Fe–Fe_3C–FeS–Mn systems, and the mechanism of the desulphurisation of iron and steel by manganese are discussed.

Steel is Desulphurised by Addition of Beryllium. W. Kroll. (Metallwirtschaft, 1934, vol. 13, Jan., pp. 21–23: Steel, 1935, vol. 96, May 27, pp. 62–63, 68). Beryllium is known to have a high affinity for oxygen; furthermore, it readily forms a sulphide, and high-sulphur steel to which beryllium has been added does not exhibit red-shortness. Many desulphurisers are effective chemically but not so effective mechanically, *i.e.*, they leave objectionable sulphide inclusions at the grain boundaries. Beryllium sulphide is fairly light and readily absorbed by any slag present, or becomes evenly distributed throughout the metal. By its use phosphorus and arsenic may also be largely removed.

Slag Control in the Making of Iron and Steel. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 625). In October, 1934, a symposium on slag control was held by the Iron and Steel Division of the American Institute of Mining and Metallurgical Engineers. The following papers which were read and discussed are presented in condensed form: Slag Control and the Blast-Furnace, by R. H. Sweetser;

Slag Control in Rimming Steel, by L. F. Reinartz ; Slag Control Applied to Low-Carbon Steel, by F. G. Norris ; Slag Control for Basic Open-Hearth High-Carbon Steel, by W. J. Reagan ; Slag Control for Recarburised Rail Steel, by A. P. Miller and T. S. Washburn ; Slag Control for Alloy Forging Steel, by A. L. Field ; Slag Control for Alloy Forging Steel, by R. C. Good ; Slag Control in Acid Open-Hearth Steel, by F. B. Foley ; Slag Control for Basic Electric-Furnace Steel, by H. F. Walther ; Physical Testing of Slag, by A. B. Kinzel.

The Recovery of Vanadium from Pig Iron. H. Zieler. (Stahl und Eisen, 1935, vol. 55, Sept. 5, pp. 963–964). The article reviews a number of papers published in Russia describing experiments made in that country to recover vanadium from pig iron made from native vanadium-bearing iron ores, and concludes with an account of experiments made at the Völklinger Hütte ; the pig was made from minette ore, and a basic Bessemer converter (not acid, as in the usual von Seth process) was used. The vanadium recovery was very low ; by adding sand to the converter slag it was improved. The author concludes, however, that for the economic recovery of vanadium the blast-furnace making the pig must be worked basically, and the vanadium recovery in the converter must be carried out under an acid slag, for the final recovery of the vanadium is only possible commercially from an acid slag.

Electric Melting Furnaces. M. Kauchtschischwili. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Aug. 31, pp. 1057–1062). Various modern types of arc and induction electric furnaces are illustrated and described.

On the Subject of the Operation of Electric Furnaces. M. Guédras. (Métaux, 1935, vol. 10, Jan., pp. 22–25). A critical commentary on a paper by R. Lemoine on the conduct of the operations in a basic electric furnace (Journ. I. and S.I., 1934, No. II., p. 563).

The Development of the Coreless Induction Furnace. (Metals and Alloys, 1935, vol. 6, May, pp. 119–124). The history of the development of the coreless induction furnace is traced. One of the earliest patents covering the use of high-frequency current for heating was issued to C. P. E. Schneider, the French ironmaster, but the patent appears to have been abandoned, and it was left to Northrup to develop the high-frequency furnace.

The Construction and Application of the Coreless Induction Furnace in Electric Steel Practice. F. Pölzguter. (Stahl und Eisen, 1935, vol. 55, July 18, pp. 773–779). The author shows that coreless induction furnaces of 4 tons capacity can nowadays be operated economically and with certainty. The plant and arrangement of a 4-ton furnace are described, and, with a 1-ton furnace as an example, the influence of differences in the form

of the windings on the output is demonstrated. By a comparison of the thermal balances of the 4-ton and 1-ton furnaces, the author completes the picture of the mode of operation of the former. From the metallurgical aspect, the furnaces have satisfied completely all demands in regard to the production of high-alloy steels. Owing to low losses by burning, they work particularly economically. For the making of special-steel castings—a comparatively new branch of the special steel industry—coreless induction furnaces have become indispensable.

Construction and Operation of Large Coreless Induction Furnaces. F. Badenheuer. (Stahl und Eisen, 1935, vol. 55, Aug. 1, pp. 821–825). The two coreless induction furnaces referred to in this article are installed in the electric steelworks of Fried. Krupp, Essen, and are largely used for making soft, high-alloy steels. After a brief description of a 750-kg. furnace, the author gives details of experiments made to determine the influence of the condition of the charge scrap or the " charge factor " on the power absorption of the furnace, and the overall efficiency is discussed from the point of view of the heat balance. The " charge factor " (*Füllfaktor*) is the ratio of the weight of the scrap placed in the furnace at the commencement of melting to the weight of the mass of metal which would completely fill the crucible to the top edge of the winding. The charge factor does not entirely characterise the quality of the scrap ; for instance, turnings which, by reason of their condition, could arrange themselves more easily along the lines of force would give less favourable results than other material which could not so adjust itself. Average current consumption values for normal operation in the furnace are also given. The author then gives a short account of a 4,000-kg. furnace and records the efficiencies obtained with it, and concludes with a description of the mode of working from the metallurgical standpoint. The author considers that large-size coreless induction furnaces have now become indispensable for the manufacture of special steels ; the limitation of the fields of application as compared with the arc furnace, with regard to the metallurgical possibilities of the two methods of steel production, is merely a question of expediency.

Developments in High Frequency Furnace Design. (Machinery (Lond.), 1935, vol. 46, July 25, pp. 511–512). The low power factor of high-frequency furnaces necessitates the use of condensers. It is shown that successful H.-F. condensers are now available, and that this fact, together with improvements in the design of the furnace casing, has made comparatively large furnaces possible.

SLAGS.

On the Viscosity of Acid and Basic Open-Hearth and Cupola Furnace Slags in Molten State. T. Matsukawa. (Taniguchi Foundation for the Promotion of Industrial Progress, Mar., 1935). The viscosities of acid and basic open-hearth and cupola slags were studied, using the rotating cylinder method. The viscosities of the steel slags were compared with that of molten iron. Acid open-hearth slag was found to be more viscous than basic slag, while both slags gave steep, but continuous, temperature/viscosity curves, there being no sudden change of slope due to primary crystallisation. The effect of fluorspar additions was examined. Fluorspar greatly reduces the viscosity of basic open-hearth slags, while, of the CaF_2 added, about 15–16 per cent. remains in the slag. Acid slags containing varying proportions of MnO, FeO, and MgO, were investigated. Ferrous oxide decreases the melting temperature and improves the fluidity, while MnO increases both the fluidity and melting temperature. If MnO is made to replace FeO the viscosity is increased. In basic slags MnO improves the fluidity and lowers the melting temperature ; a slag rich in MgO tends to be thin, but has a high melting temperature. It was found that during the refining period the viscosity of the slag is at first small, owing to the presence of iron ore, scale and fluorspar. As equilibrium is approached and material from the furnace banks is absorbed, the slag becomes thicker. In the case of cupola practice it was observed that the most fluid slag was obtained by adding 4 per cent. of limestone to the charge ; this slag also had the lowest melting point. When such a slag is used the cast iron produced has chemical and mechanical properties superior to those of other irons.

Slag Systems. R. Hay, J. White, and A. B. McIntosh. (Journal of the West of Scotland Iron and Steel Institute, 1935, vol. 42, Mar., pp. 99–103). The investigators record some results obtained by heating binary mixtures of MnO and Al_2O_3 (up to 60 per cent. Al_2O_3) for the purpose of determining the constitutional diagram and the nature of the constituents present. A number of melts were sectioned, polished and examined by the microscope under reflected light. Experiments were also undertaken on ternary slags of the FeO–MnO–SiO_2 type, with reference to the reducing effect of carbon. The slags were heated with carbon and the pressure of the CO evolved was recorded. The activity of the slags in respect of carbon oxidation was found to be : FeO, knebelite, fayalite and manganese silicate, in that order. Further experiments were performed, using small steel crucibles of 0·75 per cent. carbon content, to determine the degree of decarburisation which occurred, and hence the effect of the slag on the

iron carbide. FeO and fayalite were found to be the most active : MnO and rhodonite were comparatively inactive.

The Ternary System FeO–MnO–SiO_2. W. R. Maddocks. (Iron and Steel Institute, Carnegie Scholarship Memoirs, 1935, vol. 24, pp. 51–68). The ternary diagram FeO–MnO–SiO_2 has been studied by methods of thermal analysis. The melting surface of a wide range of compositions has been determined. A ternary eutectic appears at a composition of 50 per cent. of FeO, 20 per cent. of MnO and 30 per cent. of SiO_2, which melts at 1,170° C. A phase diagram of the system has also been constructed from the approximate measurement of the various constituents present in the solid melts. A brief résumé of the usefulness of the diagram in understanding acid open-hearth slags and the deoxidation of steel is given.

Templeborough Tar-Macadam Works. (Iron and Coal Trades Review, 1935, vol. 131, July 19, pp. 81–82). An account is furnished of the plant of the Slag Reduction Company, Ltd., for the production of tar-macadam from steelworks slag. The plant has a capacity of 1,000 tons a week. Slag from the tip is transferred by crane to a crusher and so to a rotary screen for sizing. It is again screened for grading, and transferred to slag hoppers. Belt conveyors and a skip hoist take the slag to a gas-fired drier, from which it passes to the mixer. The tar for the mixer is electrically heated, and is delivered to a mixer of the travelling-paddle type. The tar-macadam is taken from the mixer and stored in hoppers.

STEEL PLANTS.

The Manufacture of Steel in Open-Hearth Tilting Furnaces at the Appleby Plant of United Steel Companies, Ltd. A. Jackson. (Newport and District Metallurgical Society : Iron and Steel Industry, 1935, vol. 8, Aug., pp. 430–435). The author traces the manufacture of open-hearth steel from the mining of the North Lincolnshire ore to the entry of the ingots in the rolling-mill. The chief features of the melting shop are the 500-ton mixers, of which there are two, and the tilting furnaces, eight in number, and each of 120–300 tons capacity. A 150-ton fixed furnace is also in use. The author gives an account of the steel melting practice at the above works.

Acid-Bessemer Plant at Workington. (Iron and Coal Trades Review, 1935, vol. 131, Sept. 20, pp. 444–445). A description of the acid Bessemer steel plant at the Workington branch of the United Steel Companies, Ltd., is given. The melting plant consists of two Duncan Stewart-Demag 25-ton acid converters, each driven by two 125-H.P. electric motors. The converters are

worked alternately, being fed from a mixer of 400 tons capacity. The metal from the converters is tapped into 25-ton ladles and transported to the casting shop where ingots up to three tons in weight are cast.

Works of the Millom and Askam Hematite Iron Company, Limited. (Iron and Coal Trades Review, 1935, vol. 131, Sept. 20, pp. 446–449). The development of the Millom and Askam Works is traced, and an account of recent reconstruction is given. Two 78-ft. blast-furnaces, each with a capacity of 250 to 270 tons of pig iron per day have been erected, and more efficient fillings have been provided for the Cowper stoves. The plant also comprises a blast-furnace of comparatively modern construction which will undergo alteration, three turbo-blowers and a comprehensive gas cleaning plant. Sintering equipment designed for a daily output of 500 to 600 tons is in the process of erection.

The Irlam Works of the Lancashire Steel Corporation Limited. (Iron and Coal Trades Review, 1935, vol. 131, Sept. 20, pp. 468–471). The equipment of the above works includes three travelling cranes, capable of handling 4,000 tons a day, coking plant comprising fifty-one Becker ovens and the attendant by-products plant, four blast-furnaces, producing basic open-hearth iron, and eight basic open-hearth furnaces. The latter consist of five 50-ton and three 90-ton furnaces, and there is, in addition, a 300-ton active tilting mixer. The rolling-mill equipment includes a 36-in. two-high reversing cogging mill, serving a two-high reversing mill for medium and heavy sections and a three-high mill for light sections. A Morgan continuous rod mill is also installed.

The United Steel Companies. 1.—The Templeborough Works of Steel, Peech and Tozer. (Foundry Trade Journal, 1935, vol. 52, June 27, pp. 427–428). An account of the equipment at the above works is given. The melting department comprises fourteen 80-ton open-hearth steel furnaces, serving a 36-in. cogging mill rolling a 5-ton ingot into 6-in. blooms, followed by a continuous billet mill in which the blooms are reduced to 4-in. square. A second mill enables a reduction to 1½ in. square to be effected if required. The billets are cut up by flying shears and delivered to the hot bank. These billets are further reduced in the strip and bar mills of the United Strip and Bar Mills branch which lie adjacent.

The United Steel Companies, Limited. 2—Samuel Fox and Company, Limited. (Foundry Trade Journal, 1935, vol. 52, June 27, p. 428). Among the points of interest at the works of Samuel Fox and Co., Ltd., are two H.-F. steel melting furnaces of 2 and 5 tons capacity, and a central research department which meets the needs of The United Steel Companies, Limited.

East Hecla Works of Hadfields, Limited. (Foundry Trade Journal, 1935, vol. 52, June 27, pp. 431–432). A brief account of the various departments of the above concern is provided.

The Works of the South African Iron and Steel Industrial Corporation in Pretoria. W. Krebs. (Stahl und Eisen, 1935, vol. 55, Oct. 3, pp. 1057–1065). After touching on the history of the origin of the steel industry of South Africa, the author gives a detailed illustrated account of the composite works of the Corporation, and discusses the operating conditions and the results so far obtained.

Development of the Krupp Works at Essen. (Metallurgia, 1935, vol. 12, July, pp. 77–80). A short description is given of the development of the various branches of the firm of Fried. Krupp. A.-G.

150 Years of the Werk Königshuld. K. Jürgens. (Stahl und Eisen, 1935, vol. 55, Sept. 12, pp. 977–981). A brief historical survey is given of the Werk Königshuld of the Vereinigte Oberschlesische Hüttenwerke, A.-G., which celebrated its one hundred and fiftieth anniversary on Sept. 8, 1935.

A Century and a Quarter of Iron and Steel. (Iron Age, 1935, vol. 136, July 4, pp. 35–50). This article constitutes a short history of the progress and development of the Lukens Steel Co., U.S.A. The primitive methods in use in the early days of the American iron and steel industry are mentioned, and the establishment of the Lukens works is described; the material treated was charcoal iron, and the rolling mill was served by a water-wheel. Steam power was subsequently introduced and mild steel boiler plate was rolled. To-day the equipment constitutes a highly modern steelworks comprising open-hearth furnaces, a large rolling mill and auxiliary plant for the production of plate, sheet and similar material.

STEELWORKS MACHINERY.

Selecting Gearing for Cranes for Steel Mill Service. S. L. Crawshaw. (Steel, 1935, vol. 97, July 29, pp. 30–31). Some of the factors governing the choice of gears for service in steel mill cranes are considered. The author is of the opinion that spur or helical gears are to be preferred to the herring-bone type, since impact is thereby lessened and the necessity for precise axial alignment removed. He recommends that forgings be used where possible, and that if castings are employed these should be properly normalised. Data on the working tolerances and physical properties of gear teeth are also given.

Electro-Magnetic Devices for the Iron and Steel Industry. (Iron and Steel Industry, 1935, vol. 8, May, pp. 329–330). Certain

items of equipment, such as lifting magnets, magnetic chucks, clutches, and brakes, are briefly considered.

CASTING AND TREATMENT OF INGOTS.

Some Notes on Ingot Moulds. T. Swinden and G. R. Bolsover. (Institute of British Foundrymen, July, 1935: Iron and Coal Trades Review, 1935, vol. 131, July 12, pp. 44–46: Engineer, 1935, vol. 160, July 26, pp. 100–102). The authors discuss the effect of ingot mould composition on the life and cost of the mould and the quality of the resultant ingot. Data concerning the moulds used by the United Steel Companies, Ltd., are furnished. It is suggested that, on the basis of past experience, little is to be gained by using ingot moulds of alloy cast iron. The type of mould adopted will depend on many factors, but once this is decided upon, it is found that an optimum wall thickness is obtainable. The authors analyse the reasons for mould failure, and consider that "major cracking" is the prime cause, and that this is brought about by the inability of the mould to accommodate itself to the unequal expansion which takes place. Steel moulds have been tried, but have failed owing to the resultant severe distortion in service.

Copper Stools for Ingot Moulds.—I. C. E. Williams and H. B. Kinnear. (Metals and Alloys, 1935, vol. 6, July, pp. 169–172). In the first place, it is shown that, although copper stools have a lower melting point than those of cast iron, the greatly superior thermal conductivity of the copper not only enables stools of this material to be employed, but also lengthens the life of the mould and improves the quality of the ingot. The authors go on to discuss the method of casting (in water-cooled moulds), the type of copper used (tough-pitch copper of high conductivity) and the design of the stools. To avoid overheating of the surface of the stool a sufficient mass of copper must be employed—at least two-thirds the weight of the ingot cast. When failure takes place it is by local cracking and fissuration; hence, if a laminated design is adopted the sections which are thus affected may be replaced by new cakes of copper; in addition, if laminæ are used, thermal stresses have a less severe effect.

Feeder Head and its Importance in the Production of Steel. S. Wohlfahrt. (Jernkontorets Annaler, 1935: Stahl und Eisen, 1935, vol. 55, Sept. 26, pp. 1032–1033).

The Influence of Nozzle Refractories on Teeming Speeds of Ingots. A. Jackson. (Metallurgia, 1935, vol 12, Sept., pp. 158–160). Brief reference is first made to the defects which may arise from faulty teeming; they include lapping, cracking, shelling and "hanging crusts." The nozzle characteristics are shown to

affect the rate of teeming, this latter being also influenced by the head of steel in the ladle and its fluidity. The materials used for nozzles are critically examined, and in conclusion some notes on teeming speeds are given.

Sixth Report on the Heterogeneity of Steel Ingots. (Iron and Steel Institute, Sept., 1935, Special Report No. 9). This Report is divided into twelve sections, the first of which is introductory in character. The second deals with the examination of further ingots, including examples of rimming steel, semi-killed steel and killed steel, and of centrifugally cast steel. Section III., by J. H. Andrew, T. Raine, and S. B. Vickers, on the origin and formation of inclusions, and Section IV., a critical examination of the vacuum fusion method for the determination of the total oxygen in steel, by H. A. Sloman, contain the results of researches upon the oxygen content of steel. Section III. includes a comparison of the hot extraction and iodine solution methods of oxygen determination. The fifth section of the report is a bibliography of knowledge extant as regards the oxygen content of steels. T. Swinden and W. W. Stevenson contribute a section on the determination of nitrogen in steels, and this is followed by a bibliography on the occurrence of this element in steels, by G. C. Lloyd. In Section VIII., Swinden and Stevenson describe some experiments in which gases were passed through liquid steel and the effect on the soundness of the resulting ingots was noted. The section which follows, by J. H. Andrew and H. Elliss, gives the results of tests carried out on steel ingots melted *in vacuo* and under various gases. J. H. Andrew and E. M. Trent discuss the relations between segregation and gas formation, and describe experiments which throw light on segregation phenomena, whilst in Section XI. the influence of casting temperature upon the position of blow-holes in steel ingots of varying oxygen and carbon content is considered by C. A. Edwards, R. Higgins, M. Alexander, and D. G. Davies. The effect of mould design was also investigated. The last section constitutes a theoretical discussion by L. Northcott of the constitution of liquid and solid steel and of the influence of differential solidification on the segregation effects in steel ingots.

Morphology of the Inclusions in Siderurgical Products. A. M. Portevin and R. Castro. (Iron and Steel Institute, Sept., 1935; this Journal, p. 237).

Non-Metallic Inclusions in Steel. G. R. Bolsover. (Metallurgia, 1935, vol. 12, July, pp. 83–84). In this article the author describes a method of assessing the quality of a steel by counting the inclusions. Sections from the ingot under examination are viewed under the microscope; the inclusions revealed are compared with standard micrographs, and the section is rated on this basis.

FORGING AND ROLLING MILL PRACTICE.

REHEATING FURNACES.

Soaking Pits. G. R. McDermott. (Iron and Steel Engineer, 1935, vol. 12, June, pp. 369–371). The author compares the performance and construction of recuperative and regenerative soaking pits and pronounces in favour of the former. He points out that a recuperative pit enables more efficient combustion to be attained and requires a smaller mass of checker-work. The author also makes brief reference to the application of automatic pyrometer control to the firing of soaking pits.

Contribution on the Heating of Ingots in the Soaking Pit. W. Heiligenstaedt. (Stahl und Eisen, 1935, vol. 55, June 13, pp. 657–658). The author discusses the question: How long should an ingot stand in the open before charging into a soaking pit, in order that its temperature may become uniform and at the same time remain sufficient for the subsequent rolling operation? He presents diagrams showing the change of the average ingot temperature with time in the open and in the furnace, and also of the difference in temperature between the inside and outside of the ingot under the same conditions.

Anticipating Future Demands upon Continuous Heating Furnaces. H. T. Watts. (Steel, 1935, vol. 96, June 17, pp. 30–34, 58). The improvements which have been made to continuous heating furnaces, with the objects of securing greater output and more economical working, are recounted, and further suggestions are made. The process of heat transfer within the furnace is examined, and certain modifications in design are recommended with the object of increasing the heat transfer at the feed end of the furnace.

Forging Furnaces. A. M. Steever. (Industrial Heating, 1935, vol. 2, Mar., pp. 127–128; Apr., pp. 177–180; June, pp. 301–302). The furnaces are classified first as ferrous or non-ferrous, and then as batch or continuous furnaces. Some of the important factors which determine the design and performance of forging furnaces are next considered; these include the design of the controlling equipment, and the purpose for which the furnace is likely to be used.

FORGING, STAMPING, AND DRAWING.

1,000-Ton Hydraulic Press. (Engineering, 1935, vol. 139, May 31, p. 581). A brief description of a 1,000-ton press for work on steel plates up to 2 in. in thickness is given.

Solid Frame Forging Press. W. C. Kernahan. (Heat Treating and Forging, 1935, vol. 21, Apr., pp. 179–181). The frame of the press is in one piece of cast steel, and very stiff and rigid. The main shaft is a steel forging, and the main gear is of built-up welded steel construction.

750-Ton Hydraulic Presses Form Sheets for Solid Steel Auto Tops. (Steel, 1935, vol. 96, May 27, pp. 53–54). Two presses are described, each 36 ft. high, toggle-actuated, hydraulically operated, and capable of forming the roof of a motor car body every 10 sec. They are of the triple action type; the three actuating devices (blank holder, inner slide and die cushion) may be operated independently or in concert.

New Equipment Aids Production of Larger Auto Body Stampings. (Steel, 1935, vol. 97, Aug. 5, pp. 45–46). A description of a new 500-ton press, installed by the Murray Corp. of America, Detroit, is furnished. The press is of welded steel plate construction, of the four-point suspension type, and is 30 ft. high, the bed measuring 80 in. × 124 in. The speed is eight strokes per min., power being supplied by a 30-H.P. motor. The machine is used for the production of large quarter panels; for this work large sheets are required, and in order to economise material the blanks are built up from two sheets which are flash-welded together, the weld being subsequently trimmed and flattened. In this way excessive scrap loss is avoided.

Bronze. Its Use as a Forming and Drawing Material. J. D. Kaiser and G. K. Dreher. (Iron Age, 1935, vol. 136, July 25, pp. 22–23, 78). The material described is produced by Ampco Metal, Inc., Milwaukee, Wis., U.S.A. It is a bronze containing copper, aluminium and iron, with a Brinell hardness in the as-cast condition of up to 340. Dies made of this alloy are said to outlast dies of steel or cast iron and to require less frequent redressing; they also give a superior finish. Although not suitable for certain types of service—such as blanking and forging—dies of Ampco metal are stated to be excellent for ordinary sheet metal pressing operations.

Forging in the Far East. G. W. Motherwell. (Heat Treating and Forging, 1935, vol. 21, June, pp. 271–274). The author relates his experience in starting a new forge shop, at the Nissan Automobile Co., Japan, for the manufacture of "Datsun" motor cars. The shop is equipped with air-hammers, drop hammers, upsetters, and a forging press, in addition to the usual auxiliary machines. The equipment is entirely modern. The

operators had to be trained to do the work, since there appears to be no modern forging shop (other than the one described) in Japan, and hence there were no skilled workers. They proved highly adaptable, however, and the plant is now turning out enough forgings for 1,000 cars a month.

Manufacture of Steel Forgings. T. M. Service. (Metal Treatment, 1935, vol. 1, Summer Number, pp. 76–81 : Iron and Coal Trades Review, 1935, vol. 131, Sept. 6, pp. 347–349). In this article the quality of the steel, the heat treatment, and the technique of forging are considered in a general way. The methods of forging a number of typical articles —such as connecting rods, crankshafts, boiler drums and rotor forgings—are described.

Steel Forgings and Drop Stampings. A. B. Winder. (Iron and Steel Industry, 1935, vol. 8, May, pp. 331–334). In connection with heavy forgings, the author mentions that ingots up to 200 tons in weight may be cast, although the furnaces are usually acid open-hearths of only about 40–60 tons capacity. For high-alloy steels the electric furnace must be used, which somewhat limits the size of ingot obtainable. With large ingots segregation becomes important, and with both large and small ingots (as for drop stamping) reheating should be carefully done.

Production of Sound Forging Steel for the Drop-Stamper. N. H. Bacon. (Sheffield Trades Technical Societies : Iron and Steel Industry, 1935, vol. 8, Aug., pp. 436–439). The author traces the production of steel for drop-stamping from the charging of the furnaces to the examination of the billets.

Mechanical Forging Manipulators. (Iron and Coal Trades Review, 1935, vol. 131, Aug. 16, pp. 237–238). An account of the forging manipulator machines built by Davy Bros., Ltd., Sheffield, is presented. The manipulators are of the following types : (*a*) Floor travelling revolving type, (*b*) floor non-revolving straight-line type, (*c*) straight-line travelling bridge type, (*d*) fixed radial bridge type. The functions and handling capacities of these appliances are described in detail.

Drop Forging Practice. A. M. Steever. (Industrial Heating, 1935, vol. 2, Jan., pp. 14–18, 33 ; Feb., pp. 65–72). Among the subjects dealt with by the author are the following : Forging methods and materials ; dies and tools, and forging defects ; heating for forging, including the errors to be avoided and the furnaces used ; and the design, cleaning and inspection of forgings.

Hot Working and Heat Treatment of Stainless Steels. O. K. Parmiter. (Industrial Heating, 1935, vol. 2, Sept. pp. 461–466). The author classifies the types of stainless steels according to composition and applications and goes on to consider the forging treatment suitable for these steels. Instructions for forging the principal types of stainless steel, together with the temperature ranges at which the various operations should be carried out, are given.

Surface Cracking in Forged Shaft and Axle Journals. H. H. Ashdown. (Metal Progress, 1935, vol. 27, May, pp. 29–33). Axles and shafts subjected to severe service conditions sometimes develop large cracks, which are either longitudinal or akin to cracks produced by heavy grinding (shatter cracks). Two such failures were examined. It was found that the metal in each case was sound, and had received the correct mechanical and heat treatments. As an explanation the author suggests that the heavy skin stresses imposed upon axles, due to rapid braking and possibly inadequate lubrication, gave rise to surface rupture and the formation of cracks.

Influence of the Degree of Forging and of the Section Heat-Treated on the Mechanical Properties of Structural Steels. H. Korschan and E. Maurer. (Stahl und Eisen, 1935, vol. 55, Aug. 1, pp. 828–830). The mechanical properties of oil-quenched bars of carbon, manganese, nickel and chromium-nickel-molybdenum steels, 40 mm. square in cross-section were determined: the materials had been subjected to two different degrees of forging, namely, in the ratio of 1 : 65 and 1 : 5. The mechanical properties of the oil-quenched specimens tested in the form of 40 mm. square bars after forging to one-sixty-fifth, in comparison with the results obtained on longitudinal test-pieces removed from oil-quenched blooms of the same materials, 920 mm. in diam., which had been forged to one-fifth and then oil-quenched to yield the same tensile strength, revealed an increase of about 40 per cent. in the yield point, about 37 per cent. in the reduction of area and up to 128 per cent. in the notch toughness. In order to determine whether this improvement of the mechanical properties was due to the stronger action of the oil-quenching on sections only 40 mm. square or to the more severe forging, pieces 40 mm. square were cut out of the blooms of 920 mm. diam., forged five-fold, and were likewise quenched in oil and tested. Comparison of the results demonstrated that the increase in the mechanical properties in sections 40 mm. square was essentially due to the greater effect of heat treatment on the small section and not to the greater degree of forging.

The Increase of Deformability and the Decrease of the Ability to Fissure with Increasing Temperature. G. Tammann and W Müller. (Zeitschrift für Metallkunde, 1935, vol. 27, Aug., pp 187–189). The experiments described, which were carried out on iron and non-ferrous specimens, confirm practical experience that hot deformation requires the application of less force than cold-working, and that fissuration is then less to be feared than during cold deformation. With increasing temperature the liability to fissure and to cleavage and the friction during slip decrease. At raised temperatures types of slip may occur which are not possible at lower temperatures. More force is required

to produce deformation when the direction of slip is closely restricted than when slip can occur in all directions in the plane of slip; if the latter condition holds at raised temperatures, the deformability is favoured.

Stampings: Estimating Blanks for Drawn Parts. J. K. Olsen. (Machinist, 1935, vol. 79, June 1, pp. 264E–266E). The principles upon which the estimation of the areas of blanks for drawn parts is based are set forth, with illustrative examples.

Influence of the Dimensions of the Drawing Tool on the Results of the Deep-Drawing Test on Steel Sheet. W. Tonn and W. Püngel. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, May, pp. 511–514). The authors have carried out deep-drawing tests on steel sheets of various thicknesses in Erichsen and Guillery apparatus, of which the dimensions of the drawing plates were varied. It was found that the depth of drawing increased as the diameter of the punch or of the die and the radius of the drawing edge of the latter were increased; these relations were determined numerically. The influence of the wear of the drawing tool is so slight that even a badly worn testing machine will not give rise to inaccurate measurements.

Cold Finished Bars. J. D. Armour. (Metal Progress, 1935, vol. 27, May, pp. 43–48). Bars finished by turning and polishing and by grinding are briefly mentioned, and the production of cold-drawn bars is then dealt with. It is pointed out that the tensile strength of mild-steel bars may be greatly improved by cold-drawing, although the ductility is somewhat reduced. The improvement in the case of alloy or high-carbon steels is less marked. Machinability is a highly desirable property, generally attained by the introduction of high sulphur contents. Low-sulphur steels may be given good machining properties by suitable heat treatment, whilst the presence of high manganese contents (up to 1·60 per cent.) confers good machinability and allows the steel to be carburised without difficulty.

Evolution of Steel Shapes Processing. L. M. Waite. (Iron Age, 1935, vol. 136, July 25, pp. 18–20, 78, 80). Numerous attempts have been made to devise a machine for the extrusion of ferrous materials on a commercial basis. Hitherto, difficulties associated with the heavy wear on the dies have proved insurmountable. In the machine which is described, it is proposed to use the sudden explosive force of air pressure in a 14,000-ton extrusion press, the object being greatly to shorten the extrusion time, and hence the duration of contact between the hot material and the parts of the press; a time of 5 sec. is mentioned. The machine itself employs air and water pressure. Hydraulic pumps serve water vessels which are in turn coupled to the air containers. The water is made to force air into the air vessels; this is repeated until the air cylinders have become fully charged with compressed

air. Extrusion is performed by opening the appropriate valves, when the air suddenly expands against the water, which in turn exerts pressure on the extruding ram. Additional pneumatic and hydraulic cylinders are used to provide a return motion for the ram.

ROLLING-MILL PRACTICE.

The Influence of Certain Types of Roll Design on the Occurrence of Seams in Rolled Bars. H. Cramer. (Stahl und Eisen, 1935, vol. 55, July 25, pp. 797–805). Seams, which give rise to fissures in rolled material, can occur at all parts which can spread unrestrictedly, that is, where the section is not bounded by the pass; further, their occurrence is favoured by heavy pressure and by vertical boundaries to the section. Experiments made to discover the cause of this phenomenon, carried out with materials of which the original areas were equal, were abortive; the various systems of roll design were then examined, and the rectangle–oval series and its possibilities of causing defects—of which one lies in the deformation of the always cooler edges of the rectangular section—were investigated. In order to avoid the formation of minor seams in the last passes, tests were made with the following pass shapes and sequences: Oval upset pass, round upset pass, and round–oval and oval–oval pass sequences; the results obtained are detailed. The octagonal pass and its advantages are also considered. Finally, the oval–oval sequence developed by G. B. Lobkowitz and the results obtained with it are dealt with in detail.

Phenomena in the Roll Opening, in particular, the Position of the Flow Dividing-Line and the Magnitude of the Forward Flow during Rolling. T. Dahl. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, July, pp. 15–21). The author presents a mathematical study of the phenomena occurring within the opening between the rolls during the rolling of metals. Formulæ are given for the conditions for the material to be gripped by the rolls and to be drawn through the pass, and also for the position of the flow dividing-line (flow dividing-lines are lines of maximum pressure within the space in which deformation takes place; the particles of the material flow by plastic deformation in the direction of maximum pressure decrease) and for the magnitude of the forward flow of the material during rolling. Comparison of the calculated values with practical experimental data shows that the formulæ represent accurately at least the nature of the processes.

Composite Rolls. (Iron Age, 1935, vol. 136, Sept. 19, pp. 28–29). An English summary of a paper by H. Cramer describing the use of collared rolls. The original article appeared in Stahl

und Eisen, 1935, vol. 55, Feb. 28, pp. 235–239. (*See* Journ. I. and S.I., 1935, No. I. p. 395).

Pasopas Roll-Pressure Measuring Device. (Stahl und Eisen, 1935, vol. 55, Oct. 3, pp. 1070–1071). One end of a lever carries a small roller, which is pressed against the underside of the lower roll of the mill ; the other end of the lever actuates a membrane attached to a container connected to an autographic recorder, the variation of pressure (of the air or other fluid) caused by the movement of the membrane operating the recording apparatus. The mechanism records the flexure of the roll, and is calibrated to show this in terms of pressure.

Automatic Control for a Roughing Mill. L. A. Watson. (Electrical Engineering, 1935, vol. 54, June, pp. 656–660). Illustrated particulars are given of automatic screw-down equipment recently built for a 3-high roughing sheet-bar mill, and a semi-automatic screw-down control installed on another 3-high roughing mill.

Electric Controllers for Three-High Sheet Roughing Mill Screwdowns. J. J. Mellon. (Iron and Steel Engineer, 1935, vol. 12, July, pp. 430–432). The electrical controlling equipment for service in a three-high sheet roughing mill screwdown gear is described. Very rapid action was required, and care was taken to reduce the inertia of the screwdown motors to a minimum ; they were designed to work in series on the down motion and in parallel when operating in the reverse direction. Limit switches and safety devices were embodied in order to eliminate error in setting the rolls for the various passes.

Automatic Screwdown Control. R. M. Bayle. (Iron and Steel Engineer, 1935, vol. 12, July, pp. 425–429). A description of the electrical automatic screwdown control gear used in the production of heavy-flange H beams is furnished. A second type of control—that for a three-high sheet mill—is also dealt with.

Automatic Preset Screwdown Control. W. M. Ballenger. (Iron and Steel Engineer, 1935, vol. 12, July, pp. 434–436). Considered electrically, the Preset control consists of two Selsyn units coupled to two Thyratrons so as to form a relay. Operation of the controlling Selsyn causes a Thyratron tube to pass current, which in turn actuates the magnetic control of the screwdown motor. The latter comes into action and continues running until the Selsyn units are once more in the same phase relation, that is, until the angular displacement between the two Selsyns has been removed. A selector switch and group of controlling Selsyns enable each setting to be selected. As a refinement, a double Selsyn, with its attendant Thyratrons, may be adopted in order to provide a vernier adjustment ; in this way an accuracy of 0·1 per cent. may be obtained.

Self-Aligning Work-Rolls for 4-High Mills. J. H. Hitchcock. (Iron Age, 1935, vol. 135, May 2, pp. 37, 96: Iron and Steel Engineer, 1935, vol. 12, June, pp. 354–355). Misaligned rolls are likely to give rise to dangerous stresses in the roll neck bearings. To obviate this, self-aligning bearings have been introduced, in which each work-roll bearing is mounted between parallel wedges, so arranged that axial movement of the roll changes the alignment of the roll axis. The principle cannot be applied to 3-high mills or to 4-high reversing mills, since reversal of the motion would accentuate the maladjustment. The rolls are stated to give good service.

Modern Rolling-Mill Construction. W. W. Franklin. (Iron and Steel Industry, 1935, vol. 8, May, pp. 319–322; June, pp. 351–354). The author describes the methods of roll and roll-bearing adjustment adopted in modern plants and refers particularly to the screwdown gear and roll-changing devices. He examines a number of mill layouts and notes the advance of fully-automatic operation and the employment of individual electrically-driven rolls.

Tabulation of Applications of Anti-Friction Bearings to Roll Necks of Steel Rolling Mills in Service. (Iron and Steel Engineer, 1935, vol. 12, Sept., pp. 75–81). The information given includes the type of mill, number of stands, roll dimensions, the position of the bearings, the manufacturer, and the year in which the bearing was installed. Several hundred mills are tabulated; they are grouped according to type, *e.g.*, hot sheet, cold sheet, structural mills, &c.

Influence of Roll Neck Deflection on Bearing Life. A. H. Fraunthal. (Iron Age, 1935, vol. 135, May 2, pp. 30–33, 108: Iron and Steel Engineer, 1935, vol. 12, June, pp. 356–360). When a roll deflects under load the roll neck also deflects, with the result that if the bearing housing is not self-aligning the load is carried on the inner edge of the bearing. This often causes roller bearings to fail. To permit movement a spherical seating or a knife-edge may be employed, neither of which is satisfactory. The problem may be solved by the adoption of a roller, located between hardened plates on the centre line of the bearing, transverse to the axis of the rolls. The roller is mounted in a cage to retain lubricant. It is strong enough to withstand the full rolling pressure, and it is claimed to have given satisfactory results in service.

Roller Bearings Important in Precision Rolling-Mills. S. M. Weckstein. (Steel, 1935, vol. 96, July 1, pp. 34–35, 60). An account is given of the application of roller bearings to precision rolling-mills, and mainly to mills rolling wide strip of light gauge (in four-high mills), and continuous rod mills. The precautions to be adopted are indicated, and the necessity for the proper design of mill housings and mountings is stressed.

Roll Neck Bearings. G. Palmgren. (Iron and Steel Engineer, 1935, vol. 12, June, pp. 347–354). A survey is made of the application of the S.K.F. spherical roller bearing to roll neck. The development of designs for hot rolling-mill work is outlined; for this application a universal coupling has been devised to minimise shocks in transmitting the drive. The bearings are usually made a tight fit on the roll neck, but where this principle cannot be followed a loose fit mounting is employed; in this case the neck surfaces must be well lubricated. In the first mills provided with roller bearings, lubrication trouble was experienced; this was remedied by substituting soluble oil for the non-soluble oils and greases previously used. If roller bearings are installed the power required to operate the mill is often greatly reduced, while the absence of wear in the bearings enables precision rolling to be undertaken, and reduces the amount of scrap. Savings in these directions are frequently sufficient to offset the initial cost of roller-bearing equipment.

Maintenance of Roll Neck Bearings. H. D. Robb. (Iron and Steel Engineer, 1935, vol. 12, June, pp. 342–345). The application of roller bearings to roll necks requires the striking of a balance between the useful life to be expected from the bearing and the initial cost which the mill operator is willing to incur. Large-diameter bearings give a longer life, but are more expensive. Once the bearings are installed care should be taken to preserve their efficiency in the following ways: By refraining from applying excessive roll pressures (pressure gauges are useful in this connection); by attending to bearing defects promptly; by aligning the bearings carefully; by storing spares in a clean, dry place, and by using the correct type of lubricant. Spare bearings enable periodic inspection of the bearings in service to be undertaken, and the worn parts can be adjusted or reground if necessary.

Construction and Preservation of Roller Bearings of the Timken Type in Rolling-Mill Construction. E. H. Doughty. (Stahl und Eisen, 1935, vol. 55, June 27, pp. 695–703). The author reviews the various factors which gave an impetus to the development of tapered roller bearings, and then describes, with illustrations, their construction, the geometrical basis of their design, and their properties. Comparative tests on mills fitted with these and with plain bearings demonstrated the superiority of tapered roller bearings. Examples of the application of these bearings to rolling-mill rolls are given.

Construction and Preservation of Roller Bearings in Rolling-Mill Construction, Vereinigte Kugellagerfabriken A.-G. Type. W. Jürgensmeyer. (Stahl und Eisen, 1935, vol. 55, June 20, pp. 670–680). Roller bearings for use in hot- and cold-rolling mills are illustrated and described, and their advantages over plain bearings are dealt with in detail.

Construction and Preservation of Schöpf-Type Roller Bearings in Rolling-Mill Construction. A. Schöpf. (Stahl und Eisen, 1935, vol. 55, June 13, pp. 648–653). The author describes and illustrates the modern rolling-mill with single-spindle drive as used for rolling strip steel, sections, wire, &c., and the latest type of roller bearing for use in such mills, and points out the advantages of the single-spindle rolling-mill with Schöpf-type roller bearings.

Construction and Design of Roller Bearings in Rolling-Mill Construction, Kugelfischer Type. H. Schulz. (Stahl und Eisen, 1935, vol. 55, June 6, pp. 616–623). The types of bearing with cylindrical and conical rollers which may be used in rolling-mills are described. For heavily-loaded two-high cold-rolling mills special cylindrical roller bearings are used. The arrangement and connection of the inner ring to the roll neck, either as a tight fit or as an oil-lubricated running fit, and the method of mounting for section and plain rolls, are dealt with in detail and illustrated. Further, the arrangements for two-high cold-rolling mills and for special stands are described, and the advantages of roller bearings in the construction of rolling-mills are set out.

Steel Plant Lubrication. B. S. Burrell. (Mechanical Engineering, 1935, vol. 57, July, pp. 415–417). The author considers that greater attention should be paid to the problem of lubrication in the mill than has hitherto been the case. Reference is made to the lubrication of roll pinions, roll neck bearings, and anti-friction bearings; the need for cleanliness is particularly stressed.

A Short Account of the Theory of Lubrication. M. D. Hersey. (Journal of the Franklin Institute, 1935, vol. 219, June, pp. 677–702; vol. 220, July, pp. 93–119; Aug., pp. 187–214; Sept., pp. 305–331). Viscosity and friction; film thickness, film pressure and load capacity; dimensional theory, with applications and temperature rise and temperature distribution in bearings are discussed.

The Laboratory Testing of Lubricants. F. J. Slee. (Chemistry and Industry, 1935, vol. 54, Sept. 6, pp. 809–814). The author deals with the significance of the common laboratory tests and the selection of lubricants for specific purposes.

The Investigation of Lubricating Oils by means of Oil-Testing Machines. T. Rabinovitsch. (Montanistische Rundschau, 1935, vol. 27, Sept. 16). Very brief details are given of a number of oil-testing machines.

Diagrammatic Determination of the Optimum Charge Weight of Billets. H. Müller. (Stahl und Eisen, 1935, vol. 55, Aug. 29, p. 935). A nomogram is presented, by means of which it is possible to determine the most suitable weight of billet to use in order that the resulting bar may be cut to the lengths required by the order book with a minimum of waste.

Increasing the Producing and Selling Possibilities of a Steelworks by Installing New Rolling-Mill Plants. (Demag News, 1935, vol. 9, July, pp. 21–29). A number of mills recently erected by the Demag concern are described. They include a two-high reversing blooming mill in which the pinion couplings to the rolls are of the universal joint type; a continuous billet mill; and a mill for wire rods and light sections. The blooming mill reduces blooms from 230 × 190 mm. to 100 × 80 mm., the continuous billet mill reduces billets 100 × 80 mm. to bars 45 mm. square, and the wire and light section mill (consisting of a five-stand continuous roughing set, a five-stand staggered train and an eight-stand finishing set) produces angles with 20 to 50 mm. sides, flats 30 to 60 mm. wide and wire down to 5 mm. in diam.

Increasing the Production and Sales Possibilities of a Metallurgical Works by Means of New Rolling-Mill Installations. W. Nieten. (Revue de Métallurgie, Mémoires, 1935, vol. 32, July, pp. 273–281). An illustrated description is given of a rolling-mill for small steel sections and wire, built by Demag for a metallurgical works in Western Europe.

New Small-Section Mill in Staggered Arrangement. E. Weber. (Stahl und Eisen, 1935, vol. 55, Sept. 26, pp. 1029–1031). An illustrated description is given of a new small-section mill with ten roll stands, of which the first six are arranged continuously and the remaining four are placed in a staggered formation. Details are given of the furnaces, mills and auxiliary plant, together with notes on the rolling programme and the output of the mill.

Modern Steel Rod and Wire Rolling-Mill. L. Wegmann. (Stahl und Eisen, 1935, vol. 55, July 11, pp. 754–761). In order to keep pace with increased outputs, the Société Anonyme des Aciéres de Micheville have found it necessary to extend their rolling-mill plant. The new ingot mill and continuous billet mill have already been described (Stahl und Eisen, 1934, vol. 54, p. 931; 1935, vol. 55, p. 280); the present article gives an illustrated description of the small-bar and wire mills that have been installed. Data are presented of the components of the mills and their drives as well as of all appliances for speeding-up, cheapening and assuring the accuracy of the rolling process; some units, such as the cooling bed and the handling arrangements, are dealt with in greater detail.

Semi-Continuous Wire Mill in Japan. E. Kästel. (Stahl und Eisen, 1935, vol. 55, Aug. 1, pp. 831–833). Illustrations and details are given of the semi-continuous wire mill installed at the works of the Kobe Steel Works, Japan, built by the Fried. Krupp Grusonwerk Company. The plant comprises two producer-gas-fired gravity-type heating furnaces and their ingot pushers; a roller conveyor and ingot manipulator; the first continuous

train, consisting of six two-high stands with rolls 900 mm. long, 500 mm. in diam. in the first two housings and 450 mm. in diam. in the last four ; another roller conveyor passing *via* the shears ; the second continuous train, consisting of eight two-high stands with rolls 315 mm. in diam. and 800 mm. long ; two wire rod mills ; a finishing train with six stands of rolls 250–290 mm. in diam. and 800 mm. long in the first five stands and 900 mm. long in the last ; and ten automatic Garrett reels with a transfer to a chain conveyor. Ingots weighing 180 kg., measuring 140 mm. and 120 mm. square at the thick and thin ends, respectively, are rolled.

Reversing Mill for Rolling Bright Stock. (Iron and Coal Trades Review, 1935, vol. 131, Aug. 2, p. 172). The mill is of the two-high reversing type, driven by a 40-H.P. motor, and designed to deliver strip at 120 ft. per min. The rolls run in white-metal bearings, totally enclosed, and pressure-lubricated. Coilers with collapsible drums are provided at each end.

Merchant Iron and Wire Mill. (Iron and Coal Trades Review, 1935, vol. 130, June 28, pp. 1089–1090). Illustrated particulars are given of the new rolling mill installed at the Iscor Works of the South African Iron and Steel Corporation, Ltd.

Stopping Distance and Slowing-Down Time of Rolled Bars Running out from the Mill. G. Rudski. (Stahl und Eisen, 1935, vol. 55, Aug. 22, pp. 912–913). After rolled bars have passed through the flying shears, it is preferable that they should come to rest without the application of external braking. In the design of new plant this question always arises, because on the space required for the bars to stop of their own accord and on the time involved in so stopping depend the distance to be allowed between the shears and the automatic cooling bed and the time-cycle of the transfer gear moving them from the mill run-out to the cooling bed. The author shows how to calculate the required data.

Diagrammatic Calculation of Loops. H. Müller. (Stahl und Eisen, 1935, vol. 55, Sept. 12, p. 991). A nomogram is presented by means of which the length of the loop between two stands of a rod mill can be calculated. It is based on a formula in which the length of the loop is related to the length of the rod before the first pass, its cross-section before the first and before and after the second pass, and its speeds of travel through the two stands.

New Sheet-Mill Plant at the Iscor Works, Pretoria. (Iron and Coal Trades Review, 1935, vol. 131, Sept. 6, pp. 359–361). The South African Iron and Steel Industrial Corporation now operates a sheet mill at the Iscor Works, Pretoria. The mill equipment comprises six hot mills and two cold mills, with rolls 30 in. in diam. and roll-barrel lengths up to 44 in. in the case of the hot mills and 56 in. for the cold mills. Other features are heating and annealing furnaces, pickling and galvanising plant.

Steel Mill Developments. W. H. Burr. (Iron and Steel Engineer, 1935, vol. 12, Sept., pp. 34–41). This article consists of a résumé of the major advances in plant and practice in American steel mills.

Sheet and Tube's Strip Mill in Operation. C. Longenecker. (Blast Furnace and Steel Plant, 1935, vol. 23, June, pp. 399–402). A description is given of the rolling-mill for sheet and strip installed at the plant of the Youngstown Sheet and Tube Co.

Flexibility and Economy Achievement in Modernisation of Niles Sheet Mill. F. L. Prentiss. (Iron Age, 1935, vol. 136, Aug. 15, pp. 26–28, 98–99). An account of the new equipment installed at the above plant is given. A 56-in. three-high roughing mill, reducing 16-in. wide sheet bar to No. 16 gauge sheet in five passes, a bar heating furnace and two finishing mills are among the principal items.

Rolling-Mill for Heavier-Gauge Thin Sheets and Thinner Medium Sheets of Great Length. W. Krämer. (Stahl und Eisen, 1935, vol. 55, July 4, pp. 732–736). The author first indicates the disadvantages of medium-sheet rolling-mills of present-day construction and gives the limits of thickness and length of sheet that can be rolled in a single heat and with intermediate reheating. He then describes in detail the construction and operation of new types of sheet mills designed particularly for the production of thin medium sheets and the heavier gauges of thin sheets of very great length.

The Production of Blackplate. J. S. Caswell. (Blast-Furnace and Steel Plant, 1935, vol. 23, May, pp. 317–318, 324). In this, the concluding article of a series dealing with the production of blackplate, the author outlines the system of rolling employed. The through-pack and oversheet variations produced in the five-part system are examined, and it is suggested that these should and could be reduced. (*See* Journ. I. and S.I., 1935, No. I. p. 402).

Output Supervision in Thin Sheet Rolling-Mills. K. Skroch. (Stahl und Eisen, 1935, vol. 55, May 16, pp. 544–548). After a brief description of the plants to which this paper refers, the products and the outputs, the author gives an account of an investigation of the output capacity. The results are collected in diagrams, and a nomogram is given which can be used for the estimation of the schedule output and for the supervision of the schedule times. The simple reckoning process is explained by means of examples. In conclusion, the author indicates the advantages of diagrammatic representation—for instance, by the Gantt method—for the supervision of the outputs of thin sheet rolling-mills and the possibilities of applying this principle to the calculation of prime costs.

Investigations by Statistical Analysis of the Influence of the Steel Composition on the Sticking of Thin Sheets. O. Andrieu. (Stahl und Eisen, 1935, vol. 55, Aug. 29, pp. 925–930). The principal factors governing the sticking of thin sheets are the pressure, the temperature and the composition of the material, but there are other influences such as irregularities in the roll body, the adjustment of the rolls, the rolling speed, the degree of freedom of the plates from scale, the oxide coating on the sheets and the skill of the roller. The author enumerates the principal modes of sticking, and then describes tests on basic open-hearth mild-steel sheets to show the dependence of the yield of good sheets on effects due to the sheet thickness, changes in the mode of working and the composition. It is shown that there is a decided connection between the composition and the degree of sticking; iron with the lowest proportion of foreign elements has the greatest tendency to stick; contents of carbon up to 0·11 per cent. and of manganese up to 0·45 per cent. have a slight but noticeable influence, and a silicon content of 0·05 per cent. is, in general, sufficient to ensure the maximum reduction of sticking that can be attained by variation of the composition. Composition alone, however, is not a determining factor in sticking, as mechanical effects during rolling also have an influence.

New Strip Mill in Birmingham. (Iron and Coal Trades Review, 1935, vol. 131, Sept. 6, p. 354). The mill of the "B.M." Rolling-Mills, Ltd., which is described, consists of three stands of rolls, two stands having 8-in. and one 12-in. rolls, the stands being independently driven. Strip from ¼ in. to 11 in. wide in gauges from 0·300 in. down to 0·012 in. is produced.

New Hot Strip Mill at Youngstown Sheet and Tube Company Goes into Operation. (Iron and Steel Engineer, 1935, vol. 12, May, pp. 288–289). Illustrated particulars are given of the layout of this mill.

Hot Strip Steel Rolled on Steckel-Type Mill. (Iron Age, vol. 135, June 27, pp. 12–17, 78). The production of hot strip steel by the McLouth Steel Corporation, Detroit, is described. A single-stand reversing mill of the Steckel type is employed. The slab is delivered from the slab-heating furnace to the scale-breakers and edging rolls, whence it goes to the reversing mill and is reduced to a thickness of about ½ in. For subsequent passes the coiling furnaces on either side of the mill come into operation, the strip on emerging from the rolls being taken into the adjacent coiling furnace; the mill is then reversed and the strip run through in the opposite direction. On the final pass it is led to the run-out table and unloading coiler. The coiling furnaces are maintained at 1,400°–1,700° F., and are heated by gas or oil. The strip so produced is stated to be of high quality. The mill installed by the McLouth Corporation is capable of making hot

strip steel of ¼-in. to 16-in. gauge in thickness and from 5 in. to 16 in. in width. The rate of production of 12-gauge strip 12 in. wide is 15 net tons per hr.

Precision Mills for Rolling Strip, Bars and Rods. S. M. Weckstein. (American Iron and Steel Institute, May 23, 1935: Blast Furnace and Steel Plant, 1935, vol. 23, June, pp. 390–393). Some of the problems involved in the hot- and cold-rolling of strip, bars and rods are considered, chiefly with respect to the four-high mill (for strip) and continuous and semi-continuous mills (for bars and rods).

Wide Strip Mills. J. H. Van Campen. (Iron and Steel Engineer, 1935, vol. 12, June, pp. 367–369). A typical modern wide strip mill is described. The layout is as follows: First, a horizontal two-high scale breaker, then a broadside stand (two- or four-high) with roll faces of about 96 in. This deals with the slab. A four-high roughing mill of three stands follows, equipped with vertical edgers. The finishing section consists of a two-high scale breaker succeeded by six stands of four-high rolls. The scale breakers are provided with high-pressure spray lines, and the finishing stand may be followed by a flying shear. Coils of strip are produced by coilers working in tandem. The coils go to the pickling units, and then to the cold reducing mills of three stands in tandem. The material may then be further processed, in coil form or in cut lengths.

The Influence of the Roll Diameter in the Cold-Rolling of Strip Steel. W. Lueg and A. Pomp. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1935, vol. 17, No. 5, pp. 63–76). Cold-rolling tests, in which the rolling pressure and dimensions of the rolled material were measured, were carried out on strips, measuring 30 × 2 mm. by about 0·6 m. long, in a four-high mill; three open-hearth steels of different carbon contents and a dynamo steel were used, no lubrication was applied, and the working-roll diameter was varied from 45·85 mm. to 184·80 mm. As the reduction in the pass increased so the roll pressure rose, at first rapidly and then almost linearly at heavy reductions. For a constant reduction the roll pressure increased almost proportionally with the roll diameter in the range covered. The average resistance to deformation, calculated from the roll pressure and the rolled surface, passed through a maximum at medium reductions for small roll diameters, but for large diameters it continued to increase. For small diameters the acceleration (speed gain) in rolling also passed through a maximum, which shifted towards smaller reductions as the roll diameter was decreased. From the behaviour of the acceleration the coefficient of friction was calculated to be 0·07–0·11. The spread increased with the reduction, at first slowly, and then more rapidly; it also increased almost proportionally with the roll diameter. With increasing

reduction, the efficiency of deformation (the quotient of the average of the tensile strengths before and after the pass and the resistance to deformation) passed through a minimum, which became less as the roll diameter increased ; for constant reduction, the smaller the diameter the better was the efficiency. With increasing ratio of the thickness after rolling to the roll diameter, the roll pressure and the other quantities decreased rapidly and the deformation efficiency improved. With increasing tensile strength of the material rolled the roll pressure, resistance to deformation, acceleration, coefficient of friction and spread all increased more or less, but the deformation efficiency depreciated.

Defects in Strip from Continuous Mills. T. N. Keelan. (Blast-Furnace and Steel Plant, 1935, vol. 23, July, pp. 461–463, 468). The author analyses, with examples from his own experience, the defects to which continuous strip is subject.

Diagrammatic Calculation of the Costs for a Strip Mill. H. Müller. (Stahl und Eisen, 1935, vol. 55, Sept. 5, p. 966). A nomogram for calculating the operating and total prime costs for a strip mill is reproduced.

Fifty Years of the Mannesmann Skew-Rolling Process. (Röhrenindustrie, 1935, vol. 28, Feb. 11, pp. 13–14). A very brief note of the early experiments of the Mannesmann brothers, Reinhard and Max, on the production of tubes by skew-rolling. As early as 1860 their father had attempted to produce gun-barrels by rolling in parallel rolls, but without success. The brothers apparently gained a clue to the required solution from the observation that cracks, which sometimes developed into actual holes, were frequently formed in the centre of round bars when they were polished in a smooth-rolled mill with skew rolls.

The Manufacture of Chromium-Molybdenum Steel Tubes. J. Koshewnikon. (Stal, 1934 : Röhrenindustrie, 1935, vol. 28, Feb. 11, pp. 14–16 ; Mar. 11, pp. 28–29 ; Apr. 11, pp. 38–39). The author gives a detailed description of all the processes employed in the production of tubes of chromium-molybdenum steel for constructional purposes—piercing the rounds, rolling, preparation for cold-drawing, cold-drawing, annealing, normalising, hardening and tempering. The steel most usually employed is stated to contain : Carbon 0·25–0·35, manganese 0·40–0·60, chromium 0·80–1·10, molybdenum 0·15–0·25 per cent., and sulphur and phosphorus within the usual limits ; the properties and analyses required of the finished tubes are also given.

Development in Manufacture of Seamless Tubes. G. Evans. (Metallurgia, 1935, vol. 12, May, pp. 27–28, 34). The main subject of this article is the production of steel tubes by the Foren rolling-mill. In the Foren process a solid round billet is cast, and pierced by a rotary piercer to give the required inside diameter and wall thickness. It then goes to the Foren mill. A

solid mandrel of alloy steel is inserted into the billet, and the latter is operated on by pairs of rolls, disposed along the length of the billet and set at various angles, so as to exert a kneading action on the whole circumference of the billet. Stripping rolls remove the tube from the mandrel, and the process is repeated. It is claimed that six tubes, 40 ft. long, may be produced per min.; the tubes are said to possess a good surface and grain structure.

Evolution in Seamless Tube Manufacture. G. Evans. (Metal Industry, 1935, vol. 46, May 24, pp. 555–556). The early methods of manufacture are briefly referred to, and the development of the Mannesmann, Erhardt and extrusion processes is outlined. Two modern methods, making use of rotating rolls, are also described.

New Plant for the Manufacture of Weldless Tubes by the Push Bench Process. J. Marette. (Revue de Métallurgie, Mémoires, 1935, vol. 32, June, pp. 233–244). An illustrated description is given of the weldless-tube plant of the Société Métallurgique de Montbard-Aulnoye (France); it is designed for the production of tubes in sizes ranging from 55 to 200 mm. internal diam., in plain and alloy steels, the output amounting to 150 tubes per hr., 7·5 m. long, for diameters up to 75 mm., and 120 tubes per hr. for larger sizes. The Wellman Peters push bench process is used. The billets are heated in a producer-gas-fired furnace in two stages; after being preheated in the first zone of the furnace, they are pushed out, hot-sawn, and the blanks are recharged into the second zone. Thence, the blanks pass to the piercing press, and are then formed into tubes by being pushed on a mandrel through a series of drawplates of gradually diminishing size. The tubes, still on the mandrel, enter a pair of inclined rolls, which detach them from the mandrels and also polish them, inside and out. The mandrel is removed and the tubes then pass through gauging rolls to ensure that they are truly circular and the dimensions are exact. The electric driving machinery and the power requirements are also discussed.

Investigations on the Energy Consumption in the Piercing of Billets in the Mannesmann Skew Rolling-Mill. M. Wratzky. (Stal, 1934: Röhrenindustrie, 1935, vol. 28, June 11, pp. 63–64; Aug. 11, pp. 87–89). The author presents the results of nearly three hundred determinations carried out under his direction in the skew rolling-mills of the Mariopol Tube Works, on the consumption of energy and the turning moment during the piercing of round billets for 8-, 10- and 12-in tubes. His purpose was to obtain experimentally data which could be used to determine the turning moment at the motor shaft and hence the size of the driving motor required; it also appeared desirable to obtain information regarding the connection between the current con-

sumption and the peripheral speed and arrangement of the rolls during the piercing operation.

The Rolling Down of Weldless Tubes in the Continuous Mill. F. Thiel. (Röhrenindustrie, 1935, vol. 28, Sept. 11, p. 97). The author gives a brief description of a tube rolling-mill in South Russia, with notes on its operation and output.

The Design of Electric Drives. P. M. Gallo. (Blast-Furnace and Steel Plant, 1935, vol. 23, Mar., pp. 185–187; Apr., pp. 262–265; May, pp. 326–342). These three articles constitute a mathematical discussion of the design of motors and controllers such as are used in the operation of the live rollers in a steel mill. The necessary formulæ and data for the calculation of the torque and dynamic characteristics of motors, and the electro-magnetic characteristics of controllers are presented.

Electric Driving. (Engineer, 1935, vol. 160, Aug. 23, pp. 184–185; Aug. 30, pp. 210–212; Sept. 6, pp. 234–236; Sept. 13, pp. 258–259; Sept. 20, pp. 284–286; Sept. 27, pp. 310–312). The second article in this series on electric driving deals with rolling-mill drives.

Individual Motor Drives for Run Out Tables and Coilers. F. E. Harrell and C. V. Gregory. (Iron and Steel Engineer, 1935, vol. 12, Aug., pp. 1–9). The advantages of individual motor drives to run out table rollers and coilers are stressed, and the design and selection of motors for these purposes are considered in some detail.

Selection of Equipment for D.C. Power in Rolling-Mills. C. Lynn. (Iron and Steel Engineer, 1935, vol. 12, Aug., pp. 17–20). The considerations which govern the selection of motor-generator equipment are set forth.

Driving Gear for Cold Rolling-Mills. (Engineering, 1935, vol. 140, Aug. 16, p. 179). Particulars are given of the electric drives for the cold-rolling mills at the plant of the Yniscedwyn Tinplate Co., Ltd., of Swansea. The drives are for 85 H.P. and 55 H.P. The former, which is described, comprises a double-reduction helical main gear, a worm-reducing gear for barring purposes, driving and barring motors and motor-control gear. In operation the barring motor is used to start the mill, and the main motor, on taking up the drive, automatically causes the barring motor to become disengaged and to shut down.

D.-C. Circuit Breakers for Steel Mill Service. W. Deans. (Electrical Engineering, 1935, vol. 54, June, pp. 594–598). With several large motors and generators operating in parallel in a system that is coupled very closely electrically, short-circuit currents in d.c. steel mill circuits may be as high as several hundred thousand amperes with a rate of rise of from 10 to 20 million amperes per second. Circuit breakers for this service, therefore, must have high current interrupting capacity and quick response.

In this paper the circuit breakers and associated equipment for a typical modern strip mill are described. The circuit breakers are tripped not only by overload current, but also by current having too high a rate of rise.

Steel Mill Applications of General Purpose Motors. G. A. Caldwell. (Iron and Steel Engineer, 1935, vol. 12, May, pp. 294–296). Among the applications dealt with are motors for galvanising machines, for individual roll table drives and for a continuous wire-drawing machine.

ROLLING-MILL AUXILIARY PLANT.

All Movements of Billet Gauger Controlled by Seated Operator. M. M. McCall. (Iron Age, 1935, vol. 136, Aug. 8, pp. 18–19, 36–40). The construction and operating details of a billet deseaming machine built by the Niles Tool Works Co. are given. It will handle billets 5 to 12 in. square and 5½ to 18 ft. long. A table driven by a 75-H.P. motor holds the billet at an angle of 45° and brings it against the tools, the action being that of planing. Motor-operated chucking and manipulating units enable the billets to be loaded, clamped in the machine and discharged after deseaming. Centralised control permits one operator to perform all the loading, manipulating and deseaming motions.

Torch Deseaming. W. S. Farr. (Steel, 1935, vol. 97, Sept. 2, pp. 38–40). Deseaming by means of the oxygen-cutting blowpipe may be successfully adopted if a definite but simple technique is followed. As in ordinary flame-cutting, preheating is necessary; by skilful direction of the cutting flame the oxide on a preceding cut may be removed by the succeeding one, and very little subsequent cleaning is required. Light, fast cuts are best. Where possible a bed for the billets on which the operators are to work should be available, so as to facilitate turning and handling. Flame deseaming compares favourably with chipping as far as cost, efficiency, and particularly speed are concerned.

The Descaling of Steel. (Blast-Furnace and Steel Plant, 1935, vol. 23, pp. 541–545). The layout of a spraying system for the descaling of sheets, &c., during hot rolling is described, particular attention being paid to the types of valve employed.

Spray System Facilitates Descaling of Steel. (Steel, 1935, vol. 96, June 24, pp. 43–46). During the rolling process scale forms on the surface of bars or slabs which are being worked, and it is essential that this scale be removed. Removal is best effected by water sprays. A suitable installation includes a centrifugal or reciprocating pump, a hydraulic accumulator, spray valves and nozzles. The spray valves used are described in

some detail; they are required to withstand severe service, and pressures up to 1,300 lb. per sq. in. Water hammer is apt to prove troublesome, and should be guarded against.

Straightening Rolls for Standard Gauge Rails. A. Lobeck. (Stahl und Eisen, 1935, vol. 55, Aug. 1, p. 833). The use of chill-cast straightening rolls is dealt with briefly. They are mounted on horizontal axes, three above and two below the rail to be straightened, which passes through them lying on its side. A groove is turned in the face of the rolls so as to avoid pressing on the identification marks on the rails. These rolls can be used for straightening other sections; the lower rolls will last for 35,000 tons and the upper ones for 40,000 tons of rails straightened, and they can then still be used for the other sections.

Gravity Roller Conveyors Effect Uniform Output. S. Eggleston. (Steel, 1935, vol. 96, May 20, pp. 53–54, 58). Hitherto, conveyors have had but a limited application in the steel industry; the operating conditions are severe and the handling rate is often low. However, by the adoption of a suitable design of roller, and particularly of the bearings, gravity conveyors may be effectively employed.

Automatic Sheet Measuring Device. G. W. First. (Iron and Steel Engineer, 1935, vol. 12, May, pp. 292–298). This device possesses two pointers, one of which is set to the reading corresponding to the length of sheet required; the other indicates the length of each sheet as it passes the flag switches.

Measuring Devices for Measuring the Thickness of Rolled Materials. R. Albrecht. (Stahl und Eisen, 1935, vol. 55, July 18, pp. 786–789). The modern idea of interchangeability in structural parts demands that not only finished material but also semi-finished goods shall be held within certain limits of dimensions. The author illustrates and describes a number of measuring devices by means of which the thickness of sheet, wire, strip, &c., can be measured continuously during production.

Classifier for Cold-Rolled Sheets Separates Accurate from Off Gauges. (Steel, 1935, vol. 97, Sept. 9, pp. 40–41). The classifier installed at the works of the Jones and Laughlin Steel Corp., Pittsburgh, U.S.A., is designed to grade cold-rolled strip materials according to the extent to which the thickness of the strip departs —if it does so depart—from the required tolerances. The device is a flying micrometer, electrically operated. It consists of two small rolls between which the strip passes continuously on its way to the flying shears. One roll is fixed, and the other is free to move; the movement of the free roll operates an armature between induction coils, and so varies the current in the latter, according to the alteration in the thickness of the strip. The small electrical fluctuations are made to operate Thyratron tubes coupled to relays which work the strip classifying table. This

table separates the strip into piles which are above, below or within the required thickness tolerance.

Handling Devices that Make Steel Mills Modern. N. L. Davis. (Blast-Furnace and Steel Plant, 1935, vol. 23, Aug., pp. 548–553). Among the devices which are illustrated and described are a coke screening and loading appliance, a coil up-ender, a piling machine, fork-lift electric trucks, a three-pot slag car, a conveyor for coiled steel strip, and a sheet handling truck.

Blanking and Shearing Sheet Steel. (Sheet Metal Industries, 1935, vol. 9, July, pp. 443–446). The mechanical considerations which govern the shearing and blanking operations are examined in the light of the results of work carried out by investigators in this field; the determination of blanking loads, when shear on the punch or die is present or absent, as the case may be, is then considered. A nomogram for the calculation of blanking pressures is included.

MANUFACTURE OF WIRE.

Die Stresses and Die Wear. K. B. Lewis. (Wire, 1935, vol. 10, Aug., pp. 316–317). The author criticises some statements made by M. Bonzel in "Le Trefilage de l'Acier," concerning wear on dies. He advances an explanation for ringing, contending that this is the result of abrasive action, and not evidence of compressive stress as Bonzel would suggest. Lewis is of the opinion that rapid wear consequent on the excessive use of new soap is actually due to under-lubrication brought about by the flushing action of the soap; Bonzel, on the other hand, explains the cutting of the die as being due to a transference of compressive stresses to the point of exit of the wire.

Influence of the Die, the Lubricant and the Rate of Drawing on the Power Consumption in the Drawing of Fine Steel Wire, particularly on Multiple Drawing Machines. A. Pomp and H. Heckel. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1935, vol. 17, No. 8, pp. 107–126). By means of drawing tests on two carbon steels, the influence of a number of factors on the power required for drawing was determined. With widia dies the type of lubricant was of importance; soap lubricant was best, then followed drawing fluid (a mixture of water, rye meal, beef tallow and lubricating soap), while rape oil was worst. The superiority of the soap decreased with increasing number of drafts. For each lubricant an optimum die angle was found: For soap this was about 6° to 9°, and about 9° to 15° for the other lubricants. The influence of the lubricant and the die angle decreased as the reduction per draft was increased; in this way the efficiency of the drawing process was improved. At

high drawing speeds there was a very slight reduction in the power required for drawing, but it was not of practical significance. When drawing through diamond dies, the drawing fluid proved best. Contrary to the case of widia dies, drawing fluid and rape oil caused a reduction of the power for drawing, which amounted to 36 per cent. for the drawing fluid, whereas soap lubrication increased the power required. The most suitable die angle should be smaller than that for widia dies. Multiple drawing led to a slight reduction in the power for drawing, which was attributed to the backward pull applied to the wire in this drawing process. On the basis of the relations between the total power required and the total reduction of section, it was found possible to forecast the total power that would be required in further drafts.

Economical Production of Mild Steel Wire. A. W. Fielding. (Wire Industry, 1935, vol. 2, Sept., pp. 291–295). The author contends that considerable economies may be effected in the production of mild steel wire if the following procedure is adopted. The wire rod (that produced by a continuous rod mill being particularly suitable) is annealed in pots, a reducing atmosphere being maintained by means of town's gas during cooling. This prevents oxidation and confers the correct microstructure on the wire rod. Rods prepared in this way may be drawn to 25 gauge continuously at very high speeds. By butt-welding the ends of the coils of wire time otherwise taken in threading-up is saved. Although by adopting the above methods the cost of annealing is not reduced, the cost of the cleaning operations is eliminated, and a greatly increased output is obtained.

Drawing Wire Shapes. R. Saxton. (Wire, 1935, vol. 10, June, pp. 227–229). In drawing shapes heavy initial reduction is possible, but in order to avoid fracture the excessive pressure which occurs at certain points must be balanced by a relief of pressure elsewhere. Probably one of the most difficult shapes to draw successfully is the lock coil shape used in the construction of winding ropes. The author describes the production of this type of wire. The steel contains 0·75 per cent. of carbon, and and is finished to an ultimate strength of 95–100 tons per sq. in. Considerable care is necessary in setting the dies. Similar attention is required in the production of other shapes, such as ovals, squares, and half rounds.

Alloy Steels for Wire Drawing. II.—High Speed Steel. R. Saxton. (Wire Industry, 1935, vol. 2, May, pp. 169–171). The heat treatment for the class of high-speed steels amenable to treatment by wire-drawing is indicated, and the difficulties met with in the wire-drawing operation are discussed.

Alloy Steels for Wire Drawing. III.—Nickel-Iron Alloys and Steels. R. Saxton. (Wire Industry, 1935, vol. 2, June, pp. 201–202). In drawing nickel-iron alloys, such as are used in submarine

cables and electric lead-in wires, care should be taken to remove surface defects (by grinding) as these are often deep-seated. The drawing operations follow lines usual for alloy steel wires.

Alloy Steels for Wire Drawing. IV.—Steel Wire Elements and their Properties. R. Saxton. (Wire Industry, 1935, vol. 2, July, pp. 239–240). The effects of the more important alloying elements—namely, carbon, molybdenum, tungsten, manganese, silicon, chromium and nickel—on the properties of steel wire are considered.

Alloy Steels for Wire Drawing. V.—Alloy Steels Development. R. Saxton. (Wire Industry, 1935, vol. 2, Aug., pp. 267–268). A brief survey of the development of alloy steels is given, and some of the factors which govern the successful drawing of alloy steels are noted.

Cold Work Action on Stainless Steel. R. Saxton. (Metallurgia, 1935, vol. 12, Aug., pp. 119–120). The author considers some of the points which require attention when stainless steel is subjected to wire-drawing or other cold-working operations. A typical drawing programme is presented, and suggestions are made as to the recommended procedure.

Stainless Steel Wire. J. K. Findley. (Steel, 1935, vol. 97, Aug. 26, pp. 28–30). An account of the cold-drawing department for the production of stainless wire at the works of the Ludlum Steel Co. is furnished. The material for drawing is received in the form of $\frac{1}{4}$-in. round rod. It is annealed at 1,500°–1,900° F., depending on the composition of the wire, pickled in 10 per cent. sulphuric acid to remove the scale, washed, dipped in lime water and baked. The wire is drawn to 0·15 in. in three passes and then re-annealed. It is again cold-drawn, softened and lime-coated, this process being repeated until the required reduction is attained. Final drawing (on the smaller sizes) is done by diamond dies. Bright annealing is carried out by passing the wire through a tubular furnace in which a hydrogen atmosphere is maintained. High-tensile wire for springs and wire ropes is obtained by omitting some of the annealing processes, thus introducing work-hardening. Dies for coarse wire (above 0·10 in. round) are of the high-carbon high-chromium type, while tungsten carbide or diamond dies are used for the finer gauges.

Lay-Out and Maintenance of Wire Working Plant. F Waitelegg. (Wire Industry, 1935, vol. 2, Aug., pp. 255–259). Some of the items which are of importance in the layout of wire working plant—such as the straightening and crimping devices—are dealt with.

FURTHER TREATMENT OF IRON AND STEEL.

PYROMETRY.

Pyrometers in Industry. C. E. Foster. (Wild-Barfield Heat Treatment Journal, 1935, vol. 1, June, pp. 67–68; Sept., pp. 74–79). The continuation of a series of articles dealing with industrial pyrometry. (*See* Journ. I. and S.I., 1935, No. I. p. 409).

Temperature Measurement and Pyrometers. A. Cotton. (Heat Treating and Forging, 1935, vol. 21, Apr., pp. 193–197, 203). The methods available for the measurement of temperatures are briefly discussed.

Indicating Pyrometers. M. D. Pugh. (Metal Industry, 1935, vol. 46, June 14, pp. 647–649). The importance of correct temperature control in operations such as hot galvanising, tinning, babbiting, machine soldering and die-casting is stressed, and the type of pyrometer best suited for use in these processes is considered.

The Standardisation of Partial-Radiation Pyrometers by Means of the Strip-Filament Lamp. R. Steck. (Stahl und Eisen, 1935, vol. 55, July 4, pp. 737–739). After touching on the numerous factors that tend to make the readings of partial-radiation pyrometers inaccurate and the difficulties of checking them up against lamps with the ordinary type of filament, the author describes the use of a strip-filament lamp for calibration purposes, in which a metallic strip, 3 mm. wide by 25 mm. long, replaces the usual form of filament. As the temperature of the strip falls off towards the ends, owing to conduction to the supports, a notch is provided at the midpoint and the pyrometer is sighted on this.

The Installation and Maintenance of Thermo-Electric Pyrometers. G. H. Barker. (Metallurgia, 1935, vol. 12, June, pp. 57–58; July, pp. 87–89; Aug., pp. 121–122). In the first article the fundamental principles of thermo-electric pyrometry are outlined. The mode of operation of the millivoltmeter and potentiometric types of instrument are briefly sketched, and some notes on the installation of pyrometers are furnished. Succeeding articles deal with the materials used for thermocouples and their protecting sheaths, and the installation of thermocouples and their compensating leads. Finally, the checking of thermocouples is considered.

Errors in Temperature Measurement with Thermoelements. Hints on the Practical Use of Thermoelements and Protective Tubes. H. Euler and K. Guthmann. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Aug., pp. 73–90). The authors first outline the properties required of thermoelements, their characteristics and fields of application, and then deal in detail with the principal sources of error in making temperature measurements with the couples and the means of avoiding them.

HEAT-TREATMENT EQUIPMENT.

The Modern Heat Treatment of Steel Springs. (Iron and Steel of Canada, 1935, vol. 18, May-June, pp. 37–39). The equipment for the heat treatment of springs at the plant of the B. J. Coghlin Co., Ltd., is described.

Manufacture and Heat Treatment of Cutting Tools. (Industrial Heating, 1935, vol. 2, July, pp. 353–355). The tool manufacturing and heat-treating plant at the works of the Weldon Tool Co., Cleveland, Ohio, U.S.A., is described.

Stainless Steel Strip Used for Furnace Conveyor Belts. (Steel, 1935, vol. 97, Aug. 12, p. 49). The heat-treatment furnaces at the works of the Allied Products Corp., Detroit, are described. These are of the continuous type, in which belts of 18-gauge stainless steel strip are used to convey the parts through the furnace. The belt is made in one piece, 53 ft. long, and the sides are formed by punching serrations which are then bent to an angle of 90°. The furnaces operate at 600° to 1,400° F. (tempering) and 1,600° F. (hardening).

Tool Builder Installs Complete New Heat Treating Facilities. (Steel, 1935, vol. 97, Sept. 9, pp. 46–47). A description of the reorganised heat treatment department of the Monarch Machine Tool Co., Sidney, Ohio, U.S.A., is given.

Use of Electric Furnaces in Industrial Heat Treatment.—Part III. A. G. Lobley. (Metallurgia, 1935, vol. 12, May, pp. 9–13). The author reviews the types of furnace used for bright annealing and clean hardening. Pit and bell-type furnaces are described, and reference is made to water-sealed furnaces for bright annealing in an atmosphere of steam. Continuous furnaces for the treatment of sheets, tubes and small parts are also dealt with, and furnaces for the hardening of steel parts in a protective atmosphere are briefly mentioned. (*See* Journ. I. and S.I., 1935, No. I. p. 410.)

Continuous Air Drawing of Roller Bearings and Similar Parts Using the Air Heater Method. C. F. Mayer. (Industrial Heating, 1935, vol. 2, Apr., pp. 205–207). The tempering furnaces of the

Timken Roller Bearing Co. are described. They are of the air heater type, fired with natural gas; a uniform temperature is maintained by means of air circulation and thermostatic control. Compared with electric furnace operation, the air heater equipment reveals an increase in production and reduced operating costs.

A New Bright Tempering and Nitriding Equipment. G. H. S. Grene. (Wild-Barfield Heat Treatment Journal, 1935, vol. 1, Sept., pp. 70–73). The general arrangement is shown of a pit-type furnace with circulated atmosphere for the bright annealing or tempering of metals, also a modification of this, suited to nitriding.

The Development of Controlled Atmosphere Furnaces and Photoelectric Pyrometers. H. A. Wynne. (Industrial Heating, 1935, vol. 2, Jan., pp. 25–26, 34). Electric furnaces of the bell type for coiled strip, and of the continuous mesh-belt type for continuous annealing, and in which a controlled atmosphere may be maintained, are briefly referred to. Some applications of the photo-electric pyrometer also receive attention.

Atmosphere Control in Radiant Tube. W. M. Hepburn. (Metal Progress, 1935, vol. 28, Aug., pp. 41–45, 66). The first section of this article consists of a summary, in the form of a catechism, of the ideas underlying the application of radiant tubes to furnace heating. The conditions necessary for the operation of radiant tube heaters are stated, and several of the applications are cited.

CEMENTATION AND CASE-HARDENING.

Cementation of Iron and Ferrous Alloys by Glucinum [Beryllium]. J. Laissus. (Revue de Métallurgie, Mémoires, 1935, vol. 32, July, pp. 293–301; Aug., pp. 351–360). The author first reviews the work of Fetchenko-Tchopivski and of Cazaud on the cementation of steel by beryllium, and then presents an equilibrium diagram of the iron-beryllium system, comparing and interpreting the results obtained by different workers. In order to verify the conclusions drawn from these theoretical considerations and to establish the mechanism of the diffusion of beryllium in iron, a systematic study was made of the different factors influencing the phenomenon (composition of the steel, temperature, and time). Finally, series of tests were made to determine the new properties conferred on pure iron and carbon steel as a result of their cementation by beryllium (hardness, resistance to oxidation at raised temperatures, and resistance to corrosion by the atmosphere and in a wide variety of liquid media).

Metallic Cementation (IV.).—Metallic Cementation by Means of Tin Powder. T. Kase. (Kinzoku No Kenkyu, 1935, vol. 12,

Apr. 20, pp. 210–219). The cementation of iron, copper and nickel with tin at temperatures between 300° and 1,100° was investigated. Chemical and microscopic analyses, and measurements of depth of penetration and corrosion resistance were carried out. It was found that diffusion of tin into the metals occurs above 230° and the rate of diffusion increases as the temperature rises; with iron, however, the rate of diffusion changes abruptly at the A_3 point. The relation between the increase in weight of the specimen, the depth of penetration, and the absolute temperature of cementation, may be represented by an exponential function; the relation between the degree of cementation and the elapsed time may be represented by a similar function.

Tendencies and Developments in Case-Hardening Practice. H. Toplis. (Iron and Steel Industry, 1935, vol. 8, Sept., pp. 483–486). Some of the features of present-day carburising practice are described.

Carburising with Gas. (Heat Treating and Forging, 1935, vol. 21, July, pp. 327–328). At the plant of The Electric Auto-Lite Company, Toledo, Ohio, U.S.A., mixed gases are used for the carburising operation. The articles are placed in gas-heated muffles through which flow, first, re-formed gas (gas treated in a pre-combustion unit), followed by natural gas, and then by a mixture of reformed and natural gas. The process is used for treating clutch pinions, bodies and housings.

A New Case-Hardening Process. (Machinery (Lond.), 1935, vol. 46, Sept. 5, pp. 700–701). Mention is made of the Perliton liquid carburiser which is used in ordinary salt bath furnaces.

Speedicase Rapid Case-Hardening Paste. (Wild-Barfield Heat Treatment Journal, 1935, vol. 1, June, pp. 63–64). Speedicase is a liquid case-hardening medium suitable for the production of a case of moderate thickness, the maximum penetration being 0·05 in. Data are given comparing the operating costs and procedure for case-hardening with Speedicase and with the box compound usually employed.

Case-Hardening with Rapideep. (Machinist, 1935, vol. 79, May, pp. 242E–244E). Rapideep is a carburising compound of the sodium cyanide type; it is stated to enable very close control over the case-hardening operation, and, if necessary, a considerable depth of carburisation, to be attained. The depth of case may be accurately estimated on the basis of the time of immersion in the carburising bath and the strength of the cyanide bath.

Selective Carburising of Steel. L. A. Lanning. (Industrial Heating, 1935, vol. 2, July, pp. 349–350). The methods of applying local protection to those parts of case-hardened articles which are not required to be hard are briefly reviewed.

Case-Hardening Steels. C. C. Hodgson. (Metal Treatment, 1935, vol. 1, Summer Number, pp. 82–87). The general require-

ments of case-hardened material are first examined, and the alloy case-hardening steels are classified. The author proceeds to consider the influence of manganese content, quenching conditions and general heat treatment on the properties of the core and case. Finally, on the basis of this survey, a brief indication of the suitability of the various steels for different service requirements is furnished.

Case-Hardening Alloy Steels. (Metallurgist, 1935, Oct. 25. pp. 67–68). This is an English summary of an article by E. Houdremont and H. Schrader published in Archiv für das Eisenhüttenwesen, 1935, vol. 8, Apr., pp. 445–459. (*See* Journ. I. and S.I., 1935, No. I. p. 412).

Nitrogen Case-Hardening of Steels. (Metallurgist, 1935, Aug., pp. 53–56). A critical summary of work recently carried out by S. Nishigori on the mechanism of nitrogen hardening, and published in Technology Reports of the Tôhoku Imperial University, 1935, vol. 11, No. 4, is presented. Nishigori treated nearly one hundred carbon-free iron alloys containing aluminium, chromium, molybdenum, &c., by nitriding them in the usual way. They were examined by various methods, including X-rays. On the whole, these investigations appear to confirm the previous results obtained by Jones and Morgan and published in the Iron and Steel Institute, Carnegie Scholarship Memoirs, 1932, vol. 21, p. 39, and 1933, vol. 22, p. 139. The most important results obtained by Nishigori are those relating to the X-ray analysis of the nitrided steels containing aluminium and chromium.

Nitriding of Cast Iron. V. O. Homerberg. (Industrial Heating, 1935, vol. 2, Jan., pp. 19–22, 34). The uses and physical properties of nitricastiron, when sand-cast and centrifugally cast, are detailed.

Nitrided Tool Steels. H. C. Knerr. (Iron Age, 1935, vol. 136, Sept. 26, pp. 44–45, 96). Ordinary nitriding steels are not suitable for parts subjected to high surface pressures, such as tools. However, cases of tool steels which have been improved by nitriding are known, and an alloy called Cobaltchrom–PRK is of this character. It contains 1·5 to 1·7 per cent. of carbon, 13·5 per cent. of chromium, 3·3 per cent. of cobalt, and smaller amounts of molybdenum, vanadium, nickel and silicon. When nitrided in the ordinary way it develops a Vickers hardness of 1,050 to 1,150 on the case. Examples of the successful application of this steel are quoted.

Shorterising, a New Old Method of Surface Hardening. T. W. Lippert. (Iron Age, 1935, vol. 136, Aug. 8, pp. 12–17, 95). An account of the Shorter process for surface hardening by the oxy-acetylene flame is furnished. Examples of the application of the process to plain and alloy irons and steels are cited.

HEAT TREATMENT OF IRON AND STEEL.

The Scaling of Iron and Other Metals. K. Fischbeck and F. Salzer. (Metallwirtschaft, 1935, vol. 14, Sept. 13, pp. 733–739; Sept. 20, pp. 753–758). Numerous observations are on record that in reactions between compounds which proceed without the formation of liquid or gaseous phases the reaction velocity reaches unusually high values in the neighbourhood of transformation points. In connection with researches on the reactivity of solid materials, the authors have attempted to bring the reactions of solids with gases near the transformation points into line with these observations, and have concerned themselves particularly with the oxidation (scaling) of iron in the range of the α–γ transformation. It was found here that the change in form gave rise only to a discontinuous reduction of the velocity of reaction, without any jump to abnormally high values. A similar result was obtained for the transformation of α-manganese into β-manganese. The present paper is divided into two sections. In the first, of a theoretical character, the authors discuss what happens during scaling, the influence of the composition, pressure and velocity of the gas on the rate of reaction and the formation of the individual oxide phases, the mechanism of scaling and its progress with time, the influence of transformations and of foreign elements on the reactivity of solid materials, the production of protective coatings on iron and steel, and the apparent influence of the grain-size on the reactivity of solid materials (pyrophoric metals). In the second section the authors give experimental results relating to the oxidation of iron in a current of moist carbon dioxide, the scaling of iron by steam, the oxidation of iron by nitric acid, sulphur dioxide and oxygen and of manganese by oxygen, and conclude with a note on the structure of the scale layer on iron.

Subcutaneous Effects During the Scaling of Steel. R. Griffiths. (Iron and Steel Institute, Sept., 1935: this Journal, p. 67).

Equilibrium Relationships of Fe_3O_4, Fe_2O_3 and Oxygen. J. W. Greig, E. Posnjak, H. E. Merwin and R. B. Sosman. (American Journal of Science, 1935, vol. 30, Sept., pp. 239–315). The results of a study of the equilibrium relationships of Fe_3O_4, Fe_2O_3 and oxygen over a limited range of high temperatures are presented, and the equilibrium conditions are represented graphically. These conditions are discussed. In the course of the determination of the equilibrium conditions a number of observations of a more general nature were made. They include notes on oxidation, dissociation, sintering or recrystallisation, unmixing, and the structures developed by these processes as seen under the microscope; and notes on the refractive indices and colour of hematites, on X-ray photographs of the oxides, on errors found in analyses made by the Mitscherlich method, on the effect of crucibles of

platinum and of rhodium on the charges of magnetite, on the volatilisation of platinum and rhodium in oxygen, and on the consequent effect on the thermocouples.

On the Oxidation of Iron and Steels at High Temperatures and the Structure of the Scale. K. Endo. (Kinzoku No Kenkyu, 1935, vol. 12, May 20, pp. 235–244). The oxidation of iron and steel at elevated temperatures was investigated, and the nature of the scale produced was examined by the microscope and by X-rays. The scale was found to consist of three layers: Fe_2O_3, Fe_3O_4, and a mixture of FeO and Fe_3O_4. Each layer was seen to possess a characteristic crystalline appearance. The oxygen pressure of the scale was found to be a function of the temperature, position and time of oxidation.

Observations on the Oxidation of Steel. M. Baeyertz. (American Society for Metals, Oct., 1935, Preprint No. 22). Plain carbon steels, S.A.E. 2330 steel, and a series of chromium and chromium-nickel stainless steels were oxidised at elevated temperatures, and evidence of the oxidation of both silicon and chromium in advance of the oxidation of iron was obtained. When silicon and chromium were present together in the steel, there was also evidence that the silicon was preferentially oxidised before the chromium. The stainless steels oxidised according to a general pattern in which similar oxide forms occurred in the different alloys. However, these similar oxide forms occurred at different temperatures, depending on the chromium content of the alloys. The shift of oxidation type with chromium content and temperature is consistent with the known limitation of the low-chromium alloys to low-temperature applications. The resistance of the stainless steels to oxidation at elevated temperatures appears to be due to the formation of two (possibly more) oxides high in chromium.

Contributory Effects of Furnace Atmospheres on the Grain Size of Molybdenum High-Speed Steel. A. Phillips and M. J. Weldon. (American Society for Metals, Oct., 1935, Preprint No. 25). The authors have studied the effects of several furnace atmospheres on the grain-size, temperature of incipient melting and extent of decarburisation of a molybdenum high-speed steel (molybdenum 7·05, chromium 3·93, vanadium 2·05, carbon 0·77 per cent.). The heat treatments, from the standpoints of temperature and time, covered a wide range, although particular attention was devoted to the experimental conditions approaching standard hardening practice. It was found that the most profound structural alterations are produced by oxidising atmospheres, particularly those containing an appreciable water vapour concentration. No evidence was obtained that the grain growth and other abnormal effects of oxidising atmospheres are due to "self-heating" of the steel as suggested by Gill (Lecture to

American Society for Metals, National Metals Congress, New York, 1934).

Grain-Size and its Influence on Surface Decarburisation of Steel. D. H. Rowland and C. Upthegrove. (American Society for Metals, Oct., 1935, Preprint No. 21). The authors present the results of a study of the effect of grain size on the surface decarburisation of steel by moist hydrogen. The tests carried out were confined to carbon steels ranging in carbon content from 0·71 to 1·07 per cent. The data obtained indicate that grain-size influences the rate of carbon removal from steel. The effect of time and temperature on the character and width of the ferrite band for coarse and fine grained steels was determined and the results are tabulated. Columnar grain growth above and below the γ–α change point is considered, and a theory is advanced to explain the mechanism of ferrite columnarisation below the A_1 point. Analyses of the gaseous products of decarburisation were made and the mechanism of decarburisation is explained.

Causes of a Decarbonised Outer Skin. E. F. Lake. (Heat Treating and Forging, 1935, vol. 21, June, pp. 283–284). Tool steel was heated above the transformation temperature in an electric furnace and allowed to cool slowly. It was found that the higher the temperature of heating, the greater the depth of decarburisation. The same results were obtained when a furnace atmosphere of oxygen or carbon dioxide was employed, the degree of decarburisation being greater with carbon dioxide. Hydrogen, on the other hand, did not give rise to decarburisation, but steam had the same effect as air. Carbon monoxide behaved like hydrogen. This points to decarburisation as being a process of carbon removal by oxidation. It follows that annealing should not be prolonged beyond the period necessary to effect the required grain structure alteration, neither should the annealing temperature be unnecessarily high.

Heat Treating by Forced Convection. J. R. Morrison. (Industrial Heating, 1935, vol. 2, May, pp. 245–246). The principles of forced convection heating are shortly considered.

Special Atmospheres for Heat Treatment. C. L. West. (Heat Treating and Forging, 1935, vol. 21, pp. 334–337). The applications of " Elfurno " gas atmosphere, produced by the combustion of carbonaceous gases, are referred to.

Industrial Propane Used for Bright Annealing. W. Z. Friend and F. Leckie. (Iron Age, 1935, vol. 136, July 25, pp. 30–32). Industrial propane is used at the plant of the Rotary Electric Steel Co., Detroit, for the bright annealing of steel strip. Coils of strip are stacked on bases and enclosed by light alloy covers. The atmosphere for the annealing pots is produced by the partial combustion of propane in a refractory tube filled with coarse refractory impregnated with a catalyst; 16 volumes of air to 1

of propane are used, and the correct ratio is maintained by a mixer. Water vapour is removed by surface coolers and by dryers of activated alumina. The prepared gas is led to the annealing pots, through which it circulates, finally burning on emergence. A temperature of 1,400°–1,500° F. is used, with an annealing time of 7 hr. The strip annealed in this way has a very bright, clean surface.

Grain Distortion in Metals During Heat Treatment. C. Nusbaum and N. P. Goss. (Transactions of the Society for Metals, 1935, vol. 23, Sept., pp. 621–632). A series of Laue X-ray photographs is presented which shows that during the heat treatment of metals grains may grow in such a manner as to have perfect crystallinity, as indicated by sharply-defined diffraction maxima; or they may grow in such a manner as to be more or less "distorted," as indicated by diffuse or poorly-defined diffraction maxima, generally known as "radial asterism." Some of the factors which determine the freedom from or the presence of "distorted" grain growth are (1) degree of cold-work, (2) chemical composition, (3) temperature of treatment, and (4) length of time of treatment.

Maintaining the Quality of Aero-Engine Material, with Special Reference to Heat Treatment. E. R. Gadd. (Wild-Barfield Heat Treatment Journal, 1935, vol. 1, June, pp. 58–63). It is explained that material for aero-engines is required to undergo stringent tests before being accepted. Thus, the steel ingots are examined for cleanliness and soundness, whilst the bars are subjected to fracture and Brinell tests. Valve spring wire may be given tensile, reverse bend, winding and pickling tests, besides micro-examination; the latter reveals any surface decarburisation. Case-hardened parts are carburised in a mixture of 60 per cent. wood charcoal and 40 per cent. sodium (or barium) carbonate; with each batch of material a test-piece is included. Finished aircraft components may be examined for surface defects by using a suitable etching reagent, when soft spots, cracks, inclusions, &c., are made apparent. The magnetic test may be usefully employed to reveal heat-treatment and grinding cracks.

Sheet Normalising. E. R. Mort. (Iron and Steel Industry, 1935, vol. 8, June, pp. 367–370). The relative merits of open annealing, box-annealing and normalising as grain and surface control treatments for sheet material are examined. The author then deals with the types of normalising furnace generally employed and touches on the subject of controlled atmospheres; the normalising of tinplate is also mentioned.

Heat Treatments of Cast Steels. Daubois. (Arts et Métiers, 1935, vol. 87, July, pp. 147–149). The author discusses briefly the phenomena which take place within the metal during the solidification and cooling of steel castings, records a few tests

made on two steels in the condition as cast and after various heat treatments, and indicates the conclusions to be drawn from them.

Steels for the User.—X. Heat-Treatment of Steel. R. T. Rolfe. (Iron and Steel Industry, 1935, vol. 8, Aug., pp. 449–452). Continuing this series, the author deals further with heat-treatment, particularly as applied to steel castings, and cites practical examples. The treatment of worked steel is also considered, and then follows a discussion on the range of properties obtainable by suitable heat treatment.

Steels for the User.—XI. R. T. Rolfe. (Iron and Steel Industry, 1935, vol. 8, Sept., pp. 487–490). It is pointed out that the proper heat treatment of forgings and castings of carbon steel is often essential, and in any case generally desirable; the usefulness of micrographic investigation in this connection is emphasised. Finally, some space is devoted to the subject of the heat treatment of welds.

Control of Grain Size and Structure in Large Forgings by Heat Treatment. G. J. Horvitz and G. B. Jack. (Iron Age, 1935, vol. 136, Sept. 26, pp. 48–50). The authors discuss the control of grain size in carbon steel forgings ranging in section from 6 to 20 in. A few of the more important points which must be observed during the heat treatment of this class of material are listed.

Production and Heat Treatment of Springs. J. B. Nealey. (Industrial Heating, 1935, vol. 2, Mar., pp. 135–138). This article describes the equipment and procedure for forming and heat treating coil springs and elliptic springs of all sizes.

Heating for Coiling and Toughening Springs. J. B. Nealey. (Heat Treating and Forging, 1935, vol. 21, June, pp. 295–296, 303). The production of springs at the works of the American Spiral Spring and Manufacturing Co. is described. Gas-fired furnaces and heavy mechanically-operated coiling machines constitute the bulk of the equipment in the hot-wound spring section. The oil-quenching tanks are equipped with automatic conveyors. The material for springs is heated to 1,700° F., coiled, allowed to cool, reheated to 1,550° F., oil-quenched and finally tempered at 900° F. Stock for elliptic springs is heated, nibbed and cambered, and subsequently heat-treated.

Gears Manufactured into Transmission Assembly for Quiet Operation. (Steel, 1935, vol. 96, July 1, pp. 30–33). Particulars are given of the procedure adopted at the works of the Buick Motor Co. for producing transmission gears of high accuracy. The machining, lapping and heat-treating operations are described.

A New Method of Heat Treating Rails. T. Brunner. (Iron Age, 1935, vol. 136, July 11, pp. 21–24). A special normalising process, developed by the Illinois Steel Co., for the production

of rails possessing great ductility, impact and wear resistance, is described. It is anticipated that transverse fissures in rail heads will be eliminated by the adoption of this heat treatment. The process consists of cooling the rails, after they have left the mill, until the temperature in the interior of the rail section is slightly below the " thermal critical range " and above the thermal brittle range. The rails are then reheated to a temperature just above the thermal critical range, and the ends of the rail heads are cooled by an air blast ; they are thus air-quenched and are considerably harder than the body of the rail (about 375 compared with 260 Brinell). The hardness of the rail ends can be controlled by varying the severity of the air-quenching treatment ; it gradually decreases from the ends towards the middle of the rail section. A method of determining stress contours has been devised ; it consists of sawing the rail head into diamond-shaped sections, which are then measured for distortion by means of a micrometer. In this manner the stresses in old and new rails, cooled on the hot-bed and normalised, were estimated. It is shown that the normalised rail possesses the lowest initial stress.

Accurate Hardening and Tempering. E. F. Lake. (Heat Treating and Forging, 1935, vol. 21, May, pp. 219–222). The author briefly explains the principles of hardening and tempering of steels and indicates the procedure which should be followed in order to obtain the desired results.

Hardening Tool Steels in Controlled Atmospheres. S. K. Oliver. (American Society for Metals, Apr. 15, 1935 : Steel, 1935, vol. 96, June 24, pp. 30–32, 48). It is pointed out that in order to prevent excessive scaling or decarburisation when hardening tool steels, the furnace atmosphere must be carefully controlled. However, a procedure which gives good results in one locality may fail to do so elsewhere. The author discusses this fact and presents some results obtained by the use of various atmospheres. Finally, a design of furnace which will enable the proper atmosphere for any class of tool steel to be established is outlined.

Investigation Regarding a Suitable Method for Determining the Hardening Depth for Tool Steel. G. Thalén. (Jernkontorets Annaler, 1935, vol. 119, No. 3, pp. 99–134). The author suggests that the depth of hardening may be defined as the thickness of the surface layer of the test-piece measured from the edge to the point where the structure consists of equal quantities of martensite and troostite. This depth may be ascertained with sufficient accuracy by etching the fractured surface for 10 to 15 sec. with 2 per cent. nitric acid in order to darken the core. Cylindrical test-pieces $\frac{7}{8}$ in. in diam. have been found suitable for this test. The influence of the diameter of the specimen, the condition of the surface, and the heat treatment on the hardening depth has been investigated, and the results of the depth-of-

hardness test described by the author are compared with the American conception of hardenability.

Tool Steels. A. Michel. (Métaux, 1935, vol. 10, Jan., pp. 3–21 ; Apr., pp. 88–100). The author first discusses the principles of heat treatment and its effect on carbon steels, and then deals with the practical application of these principles to the forging and heat treatment of carbon tool steels.

Testing the Quality of Unalloyed Tool Steel. E. Houdremont, H. Schrader and A. Clasen. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Sept., pp. 131–142). The authors have carried out tests to determine the behaviour of unalloyed tool steels, containing about 1 per cent. of carbon, on submission to hardening, annealing, case-hardening and decarburising treatments. The steels were made in the basic electric furnace, basic or acid open-hearth furnace, crucible, and basic or acid high-frequency furnace, from charges containing various proportions of pig, scrap, sponge iron, or carbonyl iron. The repeated hardening test proved unsuitable for examining the susceptibility to hardening cracks. Determination of the hardening limits yielded marked differences in the depth of hardening and the hardenability; in general, the greater the depth of hardening the smaller was the hardening range. On bending in the hardened condition, the fully-hardening steels quenched from higher temperatures usually behaved badly. After overheating the steels during case-hardening at 1,000° C., conclusions could be drawn as to their susceptibility to overheating from the size of the cementite network in the outermost skin zone. After annealing below A_1, marked spheroidisation of the cementite was observed only in experimental steels made from particularly pure materials. On decarburising in moist hydrogen, commercial steels showed only slight differences, which were insufficient to indicate the purity of the material, but in experimental melts of particular purity the ability to decarburise was markedly affected by the method of making the steel adopted. All commercial steels had a " normal " case-hardened structure ; abnormality occurred only in materials of great purity. The occurrence of an abnormal structure in the McQuaid-Ehn test could not be correlated with the oxygen content, but it did have some relation to the purity of the steel, in so far as this brought about more rapid spheroidisation with less hysteresis and a comparatively high A_1 transformation temperature.

Experiments with Low-Tungsten Molybdenum High-Speed Steel. H. Pohl, H. Pollack and R. Scherer. (Stahl und Eisen, 1935, vol. 55, Sept. 19, pp. 1001–1005). The article is in three parts, each by one of the authors named and in the order given ; each investigator worked independently. Tests with steels containing approximately chromium 4, molybdenum 8, tungsten 2 and vanadium 1 per cent. showed that, in comparison with ordinary

high-speed steel (tungsten 18, chromium 4, vanadium 1 per cent.), melting presented no difficulties, but that the hot-working and heat treatment required greater care on account of decarburisation and volatilisation of molybdenum. The best quenching temperature was found to be 1,200°–1,220° C. (much lower than that for ordinary high-speed steels) and the best tempering temperature 540°–560° C. (somewhat lower). The serviceability of the two steels, as shown by drilling and machining tests, was about the same, though that of the molybdenum steel was very variable.

New Method of Hardening High-Speed Steel Hack-Saw Blades. J. Trickett. (Metal Treatment, 1935, vol. 1, Summer Number, pp. 100–101). The proper hardening of hack-saw blades is attended by some difficulty, since although the blade must be quite hard the ends are required to be soft. A system of hardening is described in which the blades are clamped at each end and heated by an electric current which is passed through while the blades are under slight tension. The expansion of the blade is made to operate a pointer which thus indicates the temperature reached. The blades are allowed to cool in air. Since the clamps on the ends of the blades exert a cooling effect no subsequent annealing operation is necessary.

Field Drawing of Roll Neck Bearings. S. F. Keener. (Iron and Steel Engineer, 1935, vol. 12, June, pp. 345–346). The procedure adopted in the stress-relief of roll neck bearings in service is outlined. The roller bearings are treated in an oil bath at a temperature of 325°–350° F. Since the bearings are from 18 to 53 in. in diam. and up to 6 in. in section, they must be maintained at the correct drawing temperature for a considerable period to ensure thorough treatment. Electrical heating of the oil bath is found to give the best results, and it is recommended that treatment be carried out every three or six months.

The Effects of Special Elements on the Velocity of Expansion due to Martensitization during Quenching of Steel. T. Murakami and A. Hatta. (Kinzoku No Kenkyu, 1935, vol. 12, Apr., pp. 173–188). A self-recording apparatus for obtaining expansion/time curves during the quenching of steel was constructed, and curves were obtained for four carbon steels and thirty-two alloy steels (containing up to 2 per cent. of alloying element) of nearly eutectoid carbon content. The specimens were quenched in oil at temperatures between 850° and 1,100°. The velocity of expansion due to martensite formation was calculated from a formula relating the amount formed to the elapsed time. In plain carbon steel of nearly eutectoid composition martensite is formed by oil-quenching, while in carbon steels of hyper- and hypo-eutectoid composition troostite is formed. With the exception of a 1·94 per cent. cobalt steel, the alloy steels yielded martensite. The velocity and amount of expansion of vanadium steels are small, and those

of tungsten steels are large. Among the steels containing 0·5 and 1 per cent. of alloying element added, manganese steel has the largest velocity of expansion, while among those with 1·5 or 2 per cent. of special element tungsten steel has the largest velocity.

The Deformations Accompanying the Heat Treatment of Cold-Worked Metals. M. Bonzel. (Académie des Sciences, Aug. 5, 1935: Génie Civil, 1935, vol. 107, Aug. 31, pp. 204–205). The author presents the results obtained in experiments in which cylinders (of steel and other metals) were submitted to torsion in the cold and were then heated to various temperatures, the change of the degree of torsion being noted. The results are shown graphically in deformation/temperature curves, the shapes of which are somewhat complicated. Starting from room temperature, there is first a range in which the deformation is subordinated to a displacement of the equilibrium of the internal couples as a result of the increased viscosity of the superficial layers and the raising of the elastic limit of the less-worked interior. In the zone of recrystallisation, deformations appear, the direction and amplitude of which are governed by the nature of the alloy, its condition and the degree of initial cold-work. The foregoing effects are strictly irreversible, but the author has observed that the polymorphic change in ferrous metals appears on the deformation diagrams, both on cooling and on heating, with the usual temperature lag between the Ac and Ar points and to a degree which increases with increasing initial torsion. Thanks to the diversity of those effects, the problem of nullifying the deformation on heating after previous cold-deformation seems to be capable of solution in certain cases; these the author discusses.

Influence of Cold-Deformation on the Behaviour of an Austenitic Silicon-Manganese Steel during Tempering. F. Bollenrath. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Sept., pp. 155–161). A steel containing approximately carbon 2, silicon 1, and manganese 3 per cent., obtained in the austenitic condition by quenching, was cold-worked and then tempered. Expansion/temperature curves showed that carbide precipitation occurred first, followed by the γ–α transformation. As the cold-deformation was increased, the temperature ranges in which the expansion and transformation took place were depressed to lower temperatures and became narrower. In the undeformed steel, the austenite, after the carbide precipitation, persisted over a certain temperature range, not only during further heating but also during subsequent cooling; in the heavily worked metal, on the other hand, the transformation set in immediately after, or even partially during, the precipitation at a very much increased rate. The type of the precipitation passed from the principally banded

form in the undeformed metal to an exclusively globular form in the heavily worked specimens; with intermediate degrees of working, both types occurred. Contrary to the case of unalloyed steels, the greatest increase of hardness occurred during the precipitation and not at the transformation. After the precipitation the hardness varied more with the temperature than it did in the austenitic steel. The highest temper hardness was obtained after slight deformation.

Accelerating Effect of a Sinusoidal Mechanical Tension on the Tempering of a Hyper-Quenched Iron-Nickel-Chromium-Carbon Austenite. P. Chevenard and X. Waché. (Académie des Sciences, July 22, 1935: Génie Civil, 1935, vol. 107, Aug. 10, p. 139). It has long been known that mechanical treatment can accelerate, or even initiate, the decomposition of certain metallic phases which are out of equilibrium. The authors have carried out tests on an iron alloy containing carbon 0·33, nickel 36·3 and chromium 11 per cent. This alloy, consisting of an aggregate of carbide and austenite in the stable condition when cold, can be converted into a homogeneous solid solution by hyper-quenching, and returned to the aggregate condition by tempering. The effect of tempering can be followed quantitatively by thermo-magnetic analysis; the precipitation of chromium-rich carbide raises the Curie point of the austenite and renders this constituent heterogeneous, as is shown by the spreading of the magnetic transformation. Specimens were quenched in water from 1,200° C.; they were then tempered in pairs at various temperatures and for various times, one of the pair being subjected to tension, either constant or sinusoidal, while the other was not. The results are set out in the form of graphs. The accelerating effect of sinusoidal tension on the carbide precipitation is clearly shown; the influence of a tension oscillating five times per sec. between 1·75 and 6·85 kg. per sq. mm. has the same effect as increasing the tempering temperature by about 30°; it is greater than that of a constant tension of 6·85 kg. per sq. mm., and appears to increase with the frequency of oscillation. When these alloys are expected to work at high temperatures, it is necessary to stabilise them by an appropriate anneal, as the cooling following forging is equivalent to a partial hyper-quench. Failing this stabilisation, tempering occurs during service under the double action of temperature and stress. This carbide precipitation gives rise to a certain increase in the hot strength, but in certain cases it may also cause a brittleness comparable to temper-brittleness—whence rupture without noticeable deformation, and liability to attack by certain corrosive agents (owing to the heterogeneity).

Retained Austenite and its Decomposition Range in a Quenched Cobalt High-Speed Steel. C. A. Liedholm. (Transactions of the American Society for Metals, 1935, vol. 23, Sept., pp. 672–692).

The author discusses the structure of a quenched cobalt high-speed steel, and its changes during tempering. Similarities in the microstructures after certain tempering treatments seem to indicate that quenched high-speed steels contain a constituent corresponding to the martensite of carbon steels. Studies of the specific volume also indicate that a considerable amount of austenite is retained upon quenching. The structure of a quenched high-speed steel therefore seems essentially similar in nature to that of a quenched carbon steel. The martensite and austenite decompositions occur in the same order as in carbon steels. The magnitude of the change in volume upon decomposition of the austenite makes it appear possible that some age-hardening may occur simultaneously with the decomposition.

Influence of Phosphorus on Temper-Brittleness. H. Bennek. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Sept., pp. 147–153: Metallurgist, 1935, Oct. 25, pp. 72–74). In manganese, nickel, chromium, chromium-nickel and chromium-nickel-molybdenum steels, temper-brittleness can be developed by the addition of 0·03 to 1 per cent. of phosphorus, but in unalloyed steel this is not the case. Additions of 1·5 per cent. of manganese, nickel or chromium to very low phosphorus carbon steel set up little or no temper-brittleness. Improvement of the temper-brittleness by prolonged annealing at 650° is connected with an equalisation of the phosphorus segregation by diffusion; a similar effect is obtained by a diffusion annealing before heat treatment, so that the improvement is not a reversible phenomenon. Precipitation hardening due to phosphide is increased by manganese, nickel and chromium. The opinion that phosphide precipitations in the crystal segregates play some part in causing temper-brittleness in commercial steels also, therefore, seems very probable.

Temper-Brittleness. (Metallurgist, 1935, June, pp. 43–46). Some recent work by Borel and by Nagasawa is critically summarised, and attention is then directed to the results of investigations by Bischof.

Heat Treatment of Cast Iron. R. G. McElwee. (American Foundrymen's Association, Aug., 1935, Preprint No. 3). The author discusses the production of martensitic iron and indicates some specific applications. The heat treatment of special irons is also discussed, with particular reference to an iron containing approximately 2·5 per cent. of carbon, 1·5 per cent. of silicon and 0·35 per cent. of manganese. This iron is cast white, and a tensile strength of 90,000 lb. per sq. in., an elastic limit of 70,000 lb. per sq. in, with an elongation of approximately 4 per cent. and a hardness of 235 Brinell are obtained by heat treatment.

Heat Treatment of Grey, White and Malleable Cast Irons. H. Bornstein. (Industrial Heating, 1935, vol. 2, May, pp. 235–238; June, pp. 289–292). Examples of the heat treatment

employed in the production of high quality cast irons—particularly alloy irons—are furnished.

WELDING.

Power Stations for Electric Welding. Dutilleul and Taton. (Génie Civil, 1935, vol. 107, Sept. 28, pp. 290–293). By the term "power station" is meant an assembly of converter groups coupled in parallel, capable of supplying electric current to all the arc welders in a works, the latter connecting their individual circuits on to a common main. The authors discuss the electrical characteristics of such an arrangement.

Portable Resistance Welding Machine of Large Power. J. Otto. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, June 29, pp. 805–806). The construction and use of a resistance welding machine which, though portable, is quite powerful, are described. It can be used for the butt-welding of rails without removing them from the track.

Valve Control of Spot and Seam Welders. A. C. Croskell. (Welding Industry, 1935, vol. 3, Sept., pp. 272–274). For certain welding purposes very accurate control over the welding times of spot welders is required. By the use of grid-controlled rectifiers it is possible to employ very short welding times and heavy currents, and, if necessary, a high speed of welding, as in seam welding.

A New Design of Spot Welding Electrodes. G. Komovski. (Journal of the American Welding Society, 1935, vol. 14, July, pp. 14–15). It is shown that tapered electrodes are the most suitable for spot welding, and the current density distribution for such electrodes is represented graphically. A new type of electrode is described in which, to avoid deformation of the contact surface, the end of the electrode is enclosed in a protective steel sleeve.

Electric Resistance Seam Welding and Its Applications. L. A. Ferney. (Welding Industry, 1935, vol. 3, June, pp. 171–173; July, pp. 209–212; Aug., pp. 249–250). In this series of articles the author describes the theoretical and practical aspects of seam welding, reviews the materials suitable for the process and classifies the machines available. The characteristics of the latter are critically examined, whilst attention is also given to welding technique. Many applications of seam welding are cited, some useful performance figures are given, and, finally, an example of the cost of operating a seam welder is quoted.

Electric Seam Welding Machines. (Engineer, 1935, vol. 160, Aug. 23, pp. 190–192). The principles of seam welding are

explained, the flexible nature of the process is illustrated, and the methods of controlling the welding conditions are indicated. It is shown that seam welding is in the nature of repeated spot welding, and the mechanical and electrical methods of controlling the pitch and frequency of the spot welds are compared.

Welding Developments During the Year 1934–1935. G. A. Hughes. (Iron and Steel Engineer, 1935, vol. 12, Sept., pp. 65–69). In this paper the author outlines the chief advances in American welding technique during the period under survey.

Electric Welding. G. Moressée. (Revue Universelle des Mines, 1935, vol. 11, Aug., pp. 345–361). The author explains the principles of the two processes of welding, resistance and arc welding ; he goes into more detail in regard to the latter, because it is in more general use in Belgium, and he deals with the physical, chemical and metallurgical aspects of the process. He also devotes space to a consideration of testing—tests on the weldability of steels, on the filler material, reception tests on electrodes, and tests to check the manipulative ability of the welder. The article concludes with a few references to recently-erected buildings of welded construction.

Progress of Electric Welding in the Steel Industry. H. J. Bowles. (Iron and Steel Engineer, 1935, vol. 12, May, pp. 285–287). In this article some of the aspects of the progress of welding are outlined, such as the proper training of the welders, the repair of broken or worn parts, the increased welding currents now used, and the adoption of heavily coated electrodes and alloy welding rods.

Development and Application of Electrodes in Arc Welding. W. Strelow. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Sept. 7, pp. 1080–1084). After touching on some of the essential phenomena which occur during welding with bare and coated electrodes, the author deals in more detail with the electrodes themselves—bare, coated, dipped and cored ; their properties are noted and their effects on the operation of welding are indicated.

Electric Arc Welding in General Engineering. J. Orr. (Transactions of the Institution of Engineers and Shipbuilders in Scotland, 1935, vol. 78, pp. 320–329). The results of investigations on the profiles of butt joints, the residual stresses due to welding, the welding of high-tensile steels, and fatigue tests of butt joints in mild steel and high-tensile steels, are reproduced.

Present Position of the Technique of Resistance Welding. E. Rietsch. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, June 29, pp. 800–804).

Metallurgy of Oxy-Acetylene Welding of Steel. J. H. Crichett. (International Acetylene Association, 1934 : Journal of the American Welding Society, 1935, vol. 14, May, pp. 2–6). The

author describes the production of steel in the open-hearth furnace and draws an analogy between this process and the welding operation.

Specifications Relating to Weld Metals for Oxy-Acetylene Welding. (Welding Industry, 1935, vol. 3, Aug., pp. 251–254). This article is a translation of the first three sections of the French specifications relating to weld metals for oxy-acetylene welding. The subjects covered include: General preparation of welding metals; welding metals for the welding of steels; welding metal for the welding of cast irons. These specifications are to be subjected to criticism and, if necessary, modification, before being issued in their final form.

Shop and Field Welding. E. D. Arvidsson. (Welding Industry, 1935, vol. 3, June, pp. 163–166). It is pointed out that welding in the field is generally more difficult than welding in the shop; but field welding need not be less reliable, and if proper supervision, weather protection, handling facilities and welding procedure are granted, welding *in situ* should be as efficient as shop welding. The author cites examples of welds carried out with the object of minimising welding stresses, as illustrating the possibilities of field welding.

The Welding of Low Alloy Steels. J. C. Holmberg. (Journal of the American Welding Society, 1935, vol. 14, Sept., pp. 8–10). The effects of the alloying elements in low-alloy steels for welding are discussed.

Arc Welding of High Carbon and Alloy Steels. T. N. Armstrong. (American Society for Metals, 1935, Preprint No. 4). In welding high-carbon and alloy steels by the arc process, the heat from the arc causes the formation of a hardened zone in the base metal adjacent to the deposited metal. Unequal cooling strains set up in the deposit and the base metal cause cracks to occur in this heat-affected zone. Heating the base metal prior to welding prevents welding cracks, but preheating does not appear to improve the bending properties of the weld, and apparently affects the impact properties adversely.

The Welding of High Tensile Steels. K. L. Zeyen. (Stahl und Eisen, 1935, vol. 55, Aug. 22, pp. 901–906). In the welding of high-tensile steels difficulties may arise, and these are discussed. Previously it had been considered that carbon contents up to 0·35 per cent. were permissible in plain carbon steels for welding but it is shown that by welding electrically with modern heavily-coated electrodes unalloyed steels with carbon up to about 0·55 per cent. can be welded. On the other hand, in the gas welding of thin sections difficulties often arise out of the fissuring of the weld with carbon contents as low as 0·3 per cent., so that the use of such steels is often excluded. By decreasing the carbon content and increasing the manganese, high-tensile steels with

no tendency to weld-fissuring have now been developed. In addition to the disadvantage that in the welding of high-tensile steels the zones near the welds may become hard, brittle or soft, there is the further difficulty that for such steels the usual filler materials are often unsuitable. The requirement of chemical similarity of weld-metal and basis metal is correct for corrosion-resistant steels, but not for others. For these, austenitic welding rods can well be used, even for non-austenitic plain or alloy steels, with the limitation that the tensile strength of the welds will not exceed 80 kg. per sq. mm.

Stainless Steel Containers Fabricated by Welding. (Steel, 1935, vol. 97, Aug. 5, p. 49). The advantages of welded containers of stainless steel are stressed.

Krupp Special Welding Process D.R.P. 567094. E. Beckmann. (Technische Mitteilungen Krupp, 1935, vol. 3, Aug., pp. 137–142). For long it was regarded as essential that, particularly for highly-stressed parts, the weld, the transition zone and the parent metal must have the same structure, which was attained by using electrodes of the same composition as the parent metal and by heat-treating the weld. Krupp's have now departed from this ideal. They employ an electrode with a high content of nickel and chromium, so that the deposited metal has an austenitic structure—hence the name "austenitic welding." The field of application of this special welding process is wide; for boiler and machine construction it has three particular advantages: (1) Ageing-resistant steel can be used for the parent metal; (2) the electrode produces a weld with high strength, elongation and notch-toughness qualities without subsequent heat treatment, which is resistant to ageing and has the same coefficient of expansion as the parent metal; and (3) there is no need for heat treatment to avoid welding stresses. The author touches on the plate material and the electrodes used in this process, and the properties of the welds; he also refers briefly to dynamic tests, self-stresses and their diminution by plastic deformation, and tests showing the absence of electrolytic reactions between the weld and plate which might cause damage to the boiler plate. (*See* following abstracts).

The Developing of Austenitic Welding (Set Out by Reference to the Experiments Carried Out). K. Kautz. (Technische Mitteilungen Krupp, 1935, vol. 3, Aug., pp. 143–173). The author gives a very full account of the metallographic, metallurgical and technological tests and experiments made in the development of this new welding process to which the name "austenitic welding" has been given. (*See* previous abstract.) When the process was applied to boiler making, it was necessary, on theoretical grounds, that the condition of similarity of thermal expansion of parent metal and weld should be relaxed, and this required

a large number of metallurgical and practical welding tests. As, in this process, the welded material was not to be heat-treated, further experiments were needed on this account. It was found ultimately that without heat treatment the properties of the welded joint, in the heat-affected zone of the parent metal, in the weld itself, and in the transition zone, were equal to the previous maximum requirements, and that by heat treatment no further improvement could be secured. Heat treatment to reduce the stresses was also found to be unnecessary, as the high plastic and elastic deformability and the lack of notch-sensitiveness of the austenitic weld metal were extremely favourable for the diminution of the self-stresses. It is further shown that suggestions regarding the danger of self-stresses when age-resistant and tough materials are used for the plate and weld metal, are unfounded.

Can Welds of High-Quality Materials Diminish the Life of Boiler Plates? C. Carius. (Technische Mitteilungen Krupp, 1935, vol. 3, Aug., pp. 173–175). The author describes corrosion tests on welds made by the "austenitic welding" process (*see* preceding abstracts), to determine whether the difference between the compositions of the plate and weld metal might have a harmful effect. He concludes that no such danger is to be feared.

The Welding of High-Strength Steels. K. L. Zeyen. (Technische Mitteilungen Krupp, 1935, vol. 3, Aug., pp. 176–188). The author discusses some of the metallurgical and other problems encountered in the welding of high-strength steels, under the following headings: Improving the quality of steel by alloying and heat treatment; the welding of alloy steels as a new problem; the weldability of carbon steels; the development of steels unaffected by welding for thin sections; manganese steels with higher carbon contents suitable for arc-welding; undesired changes of properties in the weld zone; progress in regard to filler metal; and the limitation of the strength of "austenitic welds" and the possibilities of attaining still higher strength values.

Transition Structures in Fusion Welding with Austenitic Filler Materials. F. Rapatz and W. Hummitzsch. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, June, pp. 555–556). Austenitic steel filler materials have the advantage that they produce tough welds which are hardly affected by heating and cooling; this advantage is particularly useful with basis metal that can be hardened considerably. In the transition zone between weld and plate, martensite is usually found—occasionally ferrite also if the carbon contents of the weld and plate are low. With higher carbon contents in the basis metal, the austenite penetrates the grain boundaries. Nickel austenite produces a tougher transition zone than manganese austenite, probably because nickel,

owing to its slower rate of diffusion, forms thinner layers of transitional martensite than manganese; a further disadvantage of manganese is the greater sensitivity of manganese martensite to overheating. Chromium and molybdenum raise the weldability, but have much less effect on the structure than nickel and manganese. When welding with manganese steels, hardening and contraction cracks occur very readily, the former because of the susceptibility of manganese martensite to overheating, and the latter on account of the large thermal expansion of austenitic manganese steels. Low carbon contents in both the basis metal and the weld favour tough joints; with nickel austenite higher carbon contents can be allowed. Good joints can be made between the known wear-resisting steel with 12 per cent. of manganese and pearlitic steels even of higher carbon contents by means of austenitic fillers with low carbon contents, particularly nickel and nickel-chromium steels.

Electric Arc Welding Bronze Overlays on to Steel. C. H. Jennings. (Iron Age, 1935, vol. 136, July 4, pp. 22–25, 178). An account is given of the welding of bronze overlays on to steel rocker rings for locomotive motors. The actual process is one of brazing rather than welding; the steel is allowed to reach a dull red heat by maintaining the deposited bronze a little ahead of the arc; thus the steel is heated by conduction and does not fuse. The metallic-arc process is used, the electrodes being of phosphor-bronze. Difficulty is experienced when attempting to build up several layers owing to the extreme fluidity of the bronze; it may be necessary to use copper dams to retain the molten metal. Grooves and corners should be well radiused in order to secure adequate penetration, and the surface to be treated should be free from grease or dirt. Given these conditions successful deposits of bronze on steel may be obtained.

Welding of Machine Tools. Hoch and L. A. Ferney. (Welding Industry, 1935, vol. 3, Aug., pp. 231–234). Machine tools such as presses, shapers, lathes, &c., of welded and cast (or bolted) construction are compared, and the merits of the welded structures are emphasised.

The Adaptation of Engine Seating Structures for Welding. J. K. Johannesen. (Welding Industry, 1935, vol. 3, Sept., p. 276). A short note on the adaptation of an engine seating structure from riveted to welded construction is given.

Induction Welding as a New Means of Manufacturing Pipes. F. Owesny, jun. (Röhrenindustrie, 1935, vol. 28, May 11, pp. 49–51). The author describes the difficulties encountered and recent advances made in the use of a method of welding tubes in which the heat required for welding is generated within the metal by the agency of a powerful alternating magnetic field applied from

without; suitably placed rollers press the edges of the shaped strip together so as to close the joint.

Electric Welding Processes in the Manufacture of Pipes and the Laying of Pipe-Lines. Kürschner. (Röhrenindustrie, 1935, vol. 28, Apr. 11, pp. 41–42; June 11, pp. 65–66). The use of resistance butt-welding, electric spot and seam welding, arc welding and "arcatom" welding in the manufacture of pipes and the laying of pipe-lines and mains is described.

The First British Automatic Flash Welding Machine for Autobodies. (Welding Industry, 1935, vol. 3, Sept., pp. 275–276).

Motor Car Body Welding. (Machinery (Lond.), 1935, vol. 46, Aug. 22, pp. 625–627). The machine has been installed at the works of the Pressed Steel Co., Ltd., Cowley, Oxford, and is used for the production of saloon motor-car bodies. It is arranged to weld the two side panels and front panels in a welding time of 8 sec. Welding is entirely automatic and the only subsequent operation is the removal of the upset fin on the outside of the body.

Experiences in the End-to-End Welding of Steel Sheets. M. Duquénoy. (Génie Civil, 1935, vol. 106, May 25, pp. 513–515). The author records the results of mechanical tests made to determine whether the junction of two sheets welded together end-to-end was as strong as the original parent metal. Six types of steel sheet were tested; each sheet was divided into three, the separate pieces serving to show (*a*) the properties of the original metal, (*b*) any changes set up by laying down a layer of weld metal and then removing it by planing, and (*c*) the properties of the weld itself (in this case the piece was cut in half and welded together again, and the test-pieces were arranged to include the weld). The author's general conclusion is that end-to-end welds in sheet steel intended to withstand tensile stresses do not merit the suspicion with which they are at present regarded in France, but he pleads for further tests on other qualities of steel in order that proof may be produced that such welds can normally be produced "100 per cent. strong."

Resistance Welding Methods in the Sheet Metal Industry.—III. L. B. Hunt and C. F. Davidson. (Sheet Metal Industries, 1935, vol. 9, July, pp. 459–460). In the concluding article in this series seam welding and projection welding are considered. It is shown that seam welding is a further development of spot welding, the spot-welds being made to overlap. Projection welding enables a large number of spot-welds to be made in one operation by the use of large, flat electrodes and localised projections on the stock to be welded. (*See* Journ. I and S.I., 1935, No. I., p. 424).

Fabrication of Road Machinery by the Electric Arc Welding Process. D. D. Trimble. (Steel, 1935, vol. 96, June 10, pp. 36–39). The essential operations are set-up and tack welding,

and final assembly welding. Bare and lightly-coated electrodes are used for tacking, but the minimum amount of tacking is performed. Special fixtures, readily adaptable to changes in the design and size of the product, are employed. Pads, pins and V blocks are used for locating the various parts. One particular fixture utilises a pillar carrying a ball and socket joint, and this enables the assembly (in this case a road grader frame) to be manipulated by one man. Some of the fixtures are mounted on rollers; others are fastened to the floor. Stress-relieving is not practicable, but a technique has been adopted which minimises the magnitude of the locked-up stresses due to welding; this is mainly accomplished by the use of heavily shielded electrodes.

Electrically-Welded Malleable Cast Iron and Its After-Treatment. T. Tilemann. (Giesserei, 1935, vol. 22, Aug. 2, pp. 377–386). The author presents the results of a detailed investigation of the optimum conditions for the welding of whiteheart and blackheart malleable cast iron by the arc- and the resistance butt-welding processes, the correct treatment of the welded material after welding and the quality of the welds produced (weldability). For arc-welding suitable coated or dipped filler rods, attached to the negative pole, must be used; compositions for the rod and coating are suggested. For thicknesses of 4 to 20 mm. a welding current of 100 amp., welding voltage of 25 v., and open-circuit voltage of 60 v., are suitable. All qualities of malleable iron can be welded, the welds being dense, homogeneous and capable of being decarburised. With suitable electrodes the carbon and silicon contents of the unwelded malleable iron can be duplicated in the weld. Layer welds on malleable iron are very wear-resistant. Resistance butt-welding can be used for whiteheart up to 15 mm. thick, but for blackheart the maximum thickness is less. The properties of butt-welds are improved by light forging on the anvil followed by annealing.

Simple cooling in air or sand is adequate for arc-welds; the tensile strength and hardness are raised and the elongation, toughness and machinability are scarcely depreciated. On heat-treating the welds in an annealing furnace in sand, best values were obtained by annealing whiteheart malleable for 30 min. at 900°–950° C. and blackheart for 2 hr. at 900°. By a normal tempering treatment the original condition of the malleable iron is more or less restored, but by a double tempering the original values are almost invariably exceeded. For simple, small and thin castings, after-treatment of arc-welds is not essential, but all butt-welds must receive treatment, and the removal of the flash improves its effect. In general, the welds in whiteheart malleable were better than those in blackheart.

Electric Welding as Applied to the Fabrication of Mild Steel Structures. E. B. Nixon. (Welding Industry, 1935, vol. 3,

Aug., pp. 237–239). Some results obtained from welding with three types of electrode (unspecified) are presented. The depth of penetration, form of the deposit and type of welding set to use are briefly examined.

Welded Plate Girders of 52 m. Span in the Construction of the Rügendamm. A. Seeger. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Sept. 28, pp. 1173–1175). Brief details are given of the long welded plate girders used in the construction of the Rügendamm, near Stralsund, and the mode of carrying out the welding is described.

Some Observations upon the Use of Arc Welding in Shipbuilding and Ship Repairing. H. W. Townshend. (Welding Industry, 1935, vol. 3, Aug., pp. 258–263). The author enumerates the economic and technical considerations which govern the application of welding to ship construction and examines typical welds from the point of view of the technique adopted. The relative merits of chamfered and squared edges, and of single-run and multi-run welds are compared. Attention is also given to methods of preconstruction.

Effect of Peening on Physical Properties of Welds and Stress-Relief. O. M. Harrelson. (Journal of the American Welding Society, 1935, vol. 14, May, pp. 19–25). The peening machine consisted of an air-hammer, beneath which the work was made to travel. To determine the effect of peening on the stresses induced by welding, a plate 1½ in. thick was slotted to give a free bar, and this was cut and chamfered for welding. A strain gauge was applied and the bar was welded by the electric arc, V and double-V butt-joints being employed. The test procedure consisted of welding several test-bars; some of them were plain welded and in others the layers were cleaned and peened. The test-bars were subsequently cut out and mechanically tested; micrographs were also obtained. It is concluded that peening relieves the stresses induced by welding, and improves the structure and physical properties of the weld.

Stress Relief of Fusion Welded Pressure Vessels. D. S. Jacobus. (Journal of the American Welding Society, 1935, vol. 14, May, pp. 6–7). The author comments on the provisions of the A.S.M.E. Code regarding stress-relieving, and on the general problem of stress-relief in pressure vessels.

X-Ray and Welding. The Foundryman's Aid to Quality Steel Castings. C. M. Underwood and E. J. Ash. (American Foundrymen's Association, Aug., 1935, Preprint No. 16). The authors describe routine methods at the Naval Gun Factory for the X-ray examination of steel castings. Welding procedure for the repair of steel castings is also described.

Irradiation of very Thick Welds with Mesothorium. R. Bert-

hold and F. Stäblein. (Technische Mitteilungen Krupp, 1935, vol. 3, No. 3, pp. 127–128).

Testing of Welded Joints. (Stahl und Eisen, 1935, vol. 55, Sept. 5, pp. 953–961). A summary of the following papers, which were presented at a joint meeting of the Verein deutscher Ingenieure and the Verein deutscher Eisenhüttenleute on Apr. 27, 1935 :

K. DAEVES : Evaluation of the tests on welded joints.
G. FIEK : Experiences in the carrying-out of fatigue tests.
K. SCHÖNROCK : Status of knowledge on the fatigue testing of welded joints.
G. BIERETT : Remarks on the measurement of weld stresses.
R. KÜHNEL : Remarks on the tensile test on welded joints.
W. HOFFMANN : The tensile test on welded joints.
M. KOMERS : Utility of the free-bending test for the testing of welded joints.
F. RAPATZ : Influence of the angle of chamfer on the results of the folding test.
A. MATTING : The tensile test, experiences and knowledge.
R. MAILÄNDER : The importance of the notched-bar impact test for the testing of welded joints.
K. BAATZ : Notch-toughness of boiler welds.
G. TICHY : The testing of the weld seam in tubes of small diameter.
H. HOFF : The forge testing of welded joints.
R. BERTHOLD : Non-destructive testing of welded joints.

Design and Testing of Welds. C. W. Hamann. (Welding Industry, 1935, vol. 3, July, pp. 220–225). The author considers welding tests under three main headings : (*a*) Preliminary tests to determine suitability for welding ; (*b*) inspection of the welding operations ; (*c*) inspection of the completed unit. The first class includes chemical analysis of the electrodes and deposited weld-metal, macro- and microscopic examination of sections cut from the finished weld, and destructive tests on prepared specimens, such as butt tensile tests, cruciform tests, fatigue tests, &c. Class (*b*) includes control of the welding conditions and of the welder. Class (*c*) embraces acoustic, electrical, magnetic and X-ray testing, and visual inspection. The author deals with the above-mentioned tests and concludes with a brief reference to research tests, as typified by photo-elastic methods of stress analysis.

A Method of Testing Welded Seams. E. H. Schmuckler. (Electric Welding, Apr., 1935 : Journal of the American Welding Society, 1935, vol. 14, July, pp. 16–17). The method of testing consists of boring a conical hole in the welded seam by means of a small portable milling machine. The same device may be used to mill off the ends of welds in order to obtain a more favourable stress distribution. The use of the Trifix cramp table in assembling lattice work for welding is also described.

The Technical, Practical and Economic Values of Comparative Methods of Testing Weld Deposits. J. P. Wadling. (Institution of Welding Engineers: Welding Industry, 1935, vol. 3, June, pp. 186–194). The author considers that the following conditions must be met: The weld metal must be uniform; the specific gravity of the weld must be adequate (connoting an absence of slag inclusions or blowholes); the metal must not be hot-short. Other important factors are: A high ratio of yield to ultimate stress; sufficient toughness; ample ductility; adequate fatigue strength.

Form of Welds and Fatigue Strength of Welded Joints. H. E. Neese. (Welding Industry, 1935, vol. 3, Oct., pp. 303–308). Fatigue tests were carried out on welded joints of various types (such as butt and strapped joints), and, for the purposes of comparison, with a riveted joint. The tests were made with a vibrating bridge of 49 ft. 2 in. span, in the centre of the lower chord of which the test bar was incorporated. The test bars were subjected to static loading, and a vibrator attached to the upper chord of the bridge induced alternating stresses, so that the total load was pulsating. Butt joints yielded the best fatigue values, being superior to the riveted specimen; a gas-welded strapped butt-joint was equal to the riveted joint. On the whole, the strength of the various welded strap joints was low. The author goes on to discuss the incidence of failure in the welded joints. Defects due to improper welding of **V** and **X** joints are responsible for the failure of butt-joints; gas welding tends to give a smoother and therefore stronger joint in the case of both butt and fillet welds.

Notched-Bar Toughness of Fusion Welds at Low and High Temperatures. W. Kleinefenn. (Iron and Steel Institute, Carnegie Scholarship Memoirs, 1935, vol. 24, pp. 87–102). In continuation of earlier work on the effects of the absorption of oxygen and nitrogen in fusion welding, the steep portion of the curve of notched-bar values has been defined for unalloyed and for nickel-alloyed weld fillets. It was found that the most favourable position of the steep part of this curve was obtained from welds made with covered electrodes, whereas exceptionally bad results were given by electric welding done with bare-wire electrodes. Normalising produced a favourable effect on autogenous welds. In electrically welded specimens no definite shifting of the steep part of the curve was established as a consequence of heat treating above the A_3 point. The effect of alloying the welding rods was to increase the notch-toughness, particularly at high temperatures.

The Distribution of Hardness in Electrically and Thermit Welded Joints. E. Cotel. (Montanistische Rundschau, 1935, Sept., Stahlbau-Technik, pp. 5–7). The author has determined the microstructure and hardness at four places on the running

surface of welds in tramway track, nine made by electric welding and five by the thermit process. The micrographs and Brinell values are recorded and briefly discussed.

Handling Distortion in Maintenance Welding Problems. G. W. Hettrick. (Journal of the American Welding Society, 1935, vol. 14, Sept., pp. 11–15). Examples of the welding of fractured parts—such as a broken milling machine, press frames, and a steam hammer cylinder—are cited. The technique of repairing these articles is described and illustrated, particular stress being laid on adequate preheating of the parts to be joined, complete penetration of the joint, correct alignment of the damaged parts, and the avoidance of residual welding stresses.

Direct Measurement of Contraction of Butt Welds. O. M. Harrelson and S. B. Slack. (Journal of the American Welding Society, 1935, vol. 14, Aug., pp. 21–22). Particulars are given of apparatus for the direct measurement of contraction in butt welds, and the results are presented of tests carried out.

A Study of Porosity in Metallic Arc Welding under Specific Conditions. J. S. Johnson. (Journal of the American Welding Society, 1935, vol. 14, July, pp. 10–14). When a seam is welded in one pass, using a copper backing bar, porosity is often experienced ; this cannot be attributed to the welding rod, speed of travel, &c., since if penetration is not complete no porosity is evident. From theoretical considerations it seems likely that carbon monoxide is responsible for the blow-holes ; the chilling effect of the backing bar allows the carbon monoxide from the metallic arc, or from oxides present on the plates being welded, to be trapped in the weld metal as it freezes. Welding experiments were performed, using standard conditions, but varying backing bars, such as copper, Nevastain, and refractory brick. Rapid chilling itself was not found to be responsible for the porosity, as revealed by X-rays. However, when rust was carefully removed from the back of the plates completely sound welds were obtained. Hence it is concluded that the unsoundness is due to the reaction between the carbon in the steel and the oxygen in the scale on the lower side of the plates.

Steel Tubes for Aircraft Construction and their Welded Joints. W. Hoffmann. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Sept. 21, pp. 1145–1148). The author first deals with the development of the steels used in aircraft construction and their mechanical properties, and then discusses the occurrence in them of cracks near welds where the metal has been under the influence of the heat of the welding operation.

Weld Metal and Rolled Steel, How do they work together? R. Jensen. (Welding Industry, 1935, vol. 3, Sept., pp. 277–281). Although the parent metal may be relied on to yield sufficient ductility when a welded plate is tested transversely, and it is

sufficient that the weld should have a higher yield point than the plate, to meet the longitudinal stresses, however, it is desirable that the weld metal should have similar characteristics to those of the plate. With this in view the author compares the strength properties of various qualities of weld metal with ship-quality plate, and especially the stress-strain diagram. The ratio between the yield point and the tensile strength was found to be higher for weld metal than for rolled steel, hence it is suggested that the properties of weld metal and plate material most nearly correspond when the former possesses the same tensile strength as the latter.

The Penetration of Molten White Metals into Stressed Steels. W. E. Goodrich. (Iron and Steel Institute, Sept., 1935: this Journal, p. 43).

Cutting Structural Steel. J. H. Zimmerman. (Mechanical Engineering, 1935, vol. 57, July, pp. 423–427). The purpose of the investigation carried out by the author was to determine the effects of the various methods of cutting structural steel on the properties of the metal adjacent to the surface. These methods were: Machining, shearing, friction sawing, hand flame-cutting, and machine flame-cutting. The results of hardness, impact and cold bend tests are reproduced, together with microstructures of the cut surfaces. Flame-cutting appears to be superior to shearing or friction sawing, the toughness being unaffected or even improved, and the hardness and microstructure being less seriously disturbed.

Steel Bars and Channels are Cut with Oxy-acetylene Flame on New Machine. (Steel, 1935, vol. 97, Sept. 9, p. 47). A flame-cutting machine which is capable of cutting, longitudinally, bars and sections up to 50 ft. long, is described. A special feature of the appliance is its self-contained carriage; the oxygen and acetylene cylinders are suspended from the carriage, and move with it, thus obviating the necessity for long tubes.

MISCELLANEOUS.

Light Steel Bottles for Gaseous Motor Fuels. W. Jamm and K. Walter. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, June 22, pp. 779–785). The authors first describe types of containers suitable for the reception of gaseous fuels for the propulsion of automobiles, and then give a short account of the various methods of manufacture and the steels used. For liquefiable gases weldless and fusion-welded bottles can be used, but for highly compressed gases only weldless containers come into consideration. For economic reasons it is essential that the ratio of the weight of the bottles to that of their contents shall

be as low as possible compatible with safety. The development of these bottles in England has led to their production in medium-alloy nickel-chromium-molybdenum steels. In Germany it has been found possible to improve considerably the properties of a low-alloy nickel-chromium-molybdenum steel (sum of the three alloying elements not exceeding 1·5 per cent.); comparison with the English steels shows the properties to be equally good, despite the lower alloy content. Tests on bottles taken from regular production demonstrate their high quality.

Norbide (B_4C)—A New Abrasive. (Iron and Coal Trades Review, 1935, vol. 131, July 12, p. 54). This material (boron carbide, B_4C) is said to be second only to the diamond in hardness. It consists of 99 per cent. pure boron and carbon, and by means of great pressure and heat may be moulded successfully without binding cements. The analysis is: Total boron 78·22, silicon 0·01, iron 0·14, total carbon 21·0, undetermined 0·63 per cent. It is inert towards chemical media and resistant to oxidation. Norbide is manufactured from carbon (as coke) and boric acid. The latter is heated to render it anhydrous, and the materials are fused in an electric resistance furnace at 2,660° C.; crystalline boron carbide results. In powdered form it finds use as an abrasive; in the moulded form it resists abrasion very well, and may be used for sand-blast nozzles, extruding dies, &c.

Metal Cleaning by Tumbling and Burnishing. N. Ransohoff. (Metal Progress, 1935, vol. 28, Sept., pp. 57–60, 76). The types of machine used for tumbling and cleaning metal parts are described. Cleaning machines are used for washing the dirt off—a process which is accomplished by immersion in a detergent solution. Batch or continuous machines may be employed for this purpose. Tumbling is done in order to remove scale or sand from forgings or castings. Small, simple parts may be tumbled on themselves, but large complicated articles require the use of stars, or other tumbling material. Tumbling may be performed dry, or with the assistance of a weak cleaning solution. Tumbling and cleaning may sometimes be combined. The final smoothing of the surface is performed by burnishing.

Measuring and Specifying Surface Finish. E. J. Abbott. (Journal of the Society of Automotive Engineers: Metal Treatment, 1935, vol. 1, Summer Number, pp. 92–94, 101). Some results obtained by the use of the Profilograph (an instrument which traces over the surface with a diamond point, the movements of which are magnified and optically recorded) are discussed. The irregularities so revealed are, when drawn to true scale, gently undulating; but by compressing the horizontal scale the surface roughness may be made to appear much more pronounced. Several such curves—termed profilograms—are reproduced; it is further shown that two kinds of irregularity may be distinguished:

Roughness and waviness. The author suggests a method of rating surfaces according to the magnitude of the peaks and valleys exhibited by the profilogram, and proceeds to examine some alternative methods of determining and assessing surface finish. He considers that of these methods, visual comparison with standard specimens and the use of the electrical pick-up (gramophone needle) are the most promising; however, they should be based on, and correlated with, the profilograms.

PICKLING.

A Modern Pickling Plant for a Tinplate Mill. (Iron and Coal Trades Review, 1935, vol. 130, May 31, pp. 934–935: Sheet Metal Industries, 1935, vol. 9, July, pp. 421–423.) The plant described is arranged for continuous operation; a loop-line is used, part of which may be lowered in order to immerse the cradles containing the sheets in the pickling vats. Four such vats are employed, one containing fresh liquor (H_2SO_4), one containing spent liquor, and the others containing washing liquor. When immersed the cradles are given a rocking motion. A full working cycle requires only 45 sec.

The Bullard-Dunn Process for Descaling Metals. (Machinery (Lond.), 1935, vol. 46, July 4, pp. 426–427). In the Bullard-Dunn process the work is placed in the descaling bath (which contains sulphuric acid as the base) and made the cathode of an electric circuit. Hydrogen is evolved, and serves to remove the scale: simultaneously a protective film of a suitable metal—lead or tin—is deposited on the descaled surface. Work which is oily or greasy is subjected to electrolytic degreasing in an alkaline bath before treatment. The equipment required is similar to that used in electroplating work. The Bullard-Dunn process is stated to be highly effective and applicable to forgings, heat-treated parts, &c.

Modern Electrolytic Pickling Processes. O. Ungersböck. (Zeitschrift des Oesterreichischen Ingenieur und Architekten Vereines, 1935, vol. 87, May 17, pp. 116–117). After explaining very briefly the reactions at the electrodes when electric current is passed through them and an electrolyte into which they dip, the author gives a short account of electrolytic pickling without direct contact of the articles with the electrodes. The objects are placed between the electrodes, and as the current follows the path of least resistance, practically all of it enters one side of the article and leaves by the other. By rotating the article the advantages of the reversal of polarity are obtained. The process is more rapid than the ordinary pickling process, and does not

slow down with time. The greater plant cost is compensated for by the lower operating costs due to the shorter time.

A.C. Electric Pickling System Involves Use of Organic Addition Agent. (Steel, 1935, vol. 96, June 17, p. 55). The method described is known as the Ferrolite system and has been developed by the Ferro Enamel Corp., Cleveland, U.S.A. There is no direct connection between the work and the electric circuit, but the ionising effect of the electric current stimulates the action of the bath. To the normal inorganic acid bath an organic material (Ferrolite) is added which has a high affinity for oxides of iron. The ferrous sulphate is oxidised to ferric sulphate and this combines with the normal salt of gluconic acid, ferrous gluconate, to form ferrigluconic acid, and permits the sulphate ions to combine with the hydrogen present and form sulphuric acid. It is claimed that the process results in longer life of the pickling solution, lower acid cost, and a better finish.

The Acid Pickling of Ferrous Materials. (Wire Industry, 1935, vol. 2, July, pp. 245–247). The mechanism of pickling is briefly touched on, and attention is then directed to the use of inhibitors. These are divided into two classes, (*a*) foaming, (*b*) non-foaming. The former type prevent spraying and assist in holding up surface scum; the latter are adopted where the use of a foam is considered objectionable. Both types are equally effective.

The Use of Sodium Cyanide in a Neutralising Bath in Pickling Sheet Iron for Enamelling. E. H. Shands, G. L. Bruton and G. J. Grimes. (Journal of the American Ceramic Society, 1935, vol. 18, Oct., pp. 323–324). Sodium cyanide has been found to be effective as a neutraliser when used in equal parts with soda ash. Its principal advantages are (1) lengthened life of the solution, (2) increased speed of the neutralising reaction, (3) more thorough penetration into the pores of the steel, and (4) economy in plant operation.

Inhibitors. K. B. Lewis. (Wire and Wire Products, 1935, vol. 10, June, pp. 236–237, 243). Some results obtained by foreign investigators in the field of wire pickling are discussed, brief reference being made to the effect of hydrogen and of inhibitors.

The Prevention of Corrosion by Acids. (Chemical Age, 1935, vol. 33, Aug. 3, Metallurgical Section, pp. 9–10). A study is made of the use and effect of inhibitors during pickling. The part played by hydrogen during the pickling process is briefly referred to, and the behaviour of certain inhibitors in alkaline and acid solutions is shortly considered.

Contribution on the Diffusion of Hydrogen during the Pickling of Sheet Steel. H. Bablik. (Korrosion und Metallschutz, 1935, vol. 11, Aug., pp. 169–172). It has been found in practical pickling and in experimental tests that the heavy increase in the

diffusion of hydrogen in the iron which occurs, although the solubility of the metal and other conditions remain unchanged, is due to the presence of sulphur compounds in the pickling acid. It therefore follows that, in order that defects due to excessive hydrogen diffusion may be avoided, the pickling baths must be kept as free from sulphuretted hydrogen as possible.

Pickle Pitting by Electrolytic Potentials as Affected by Scaling Temperature. C. H. McCollam and D. L. Warrick. (American Society for Metals, Oct., 1935, Preprint No. 11). Pickle pitting of steel may be defined as a localised attack during the pickling operation, resulting in surface pits of various sizes and depths. It is frequently caused by an electrolytic couple, set up between the scale-free areas and scale formed above a certain critical temperature. This temperature varies with steels of different compositions. When scale is formed below the transition temperature, neither the difference in electrical potential nor the depth of pitting from this cause are significant. When this temperature is exceeded, the difference in potential, found to be approximately 0·6 v., results in a galvanic action between the scale and the metal, which proceeds at the expense of the metal. The magnitude of the resulting pit is a function of time and current density. The effect of mechanical scale-breaking, electrolytic pickling, and pickling inhibitors on this type of pitting is discussed.

COATING OF METALS.

(For Corrosion of Metals, see p. 520).

Influence of Protective Layers on the Life of Metals. F. N. Speller. (American Society of Mechanical Engineers, Iron and Steel Division, June, 1935: Mechanical Engineering, 1935, vol. 57, June, pp. 355–360). The author classifies corrosion as atmospheric, underwater, soil, chemical, and electrolytic, according to the environment. He goes on to discuss the formation of surface films of a corrosion-resistant type, and the value of oxide coatings, paints and lacquers, and of coatings particularly suitable for underground protection.

Metallic Coatings and Protective Media. S. Robson and P. S. Lewis. (Chemistry and Industry, 1935, vol. 54, June 28, pp. 605–616). The authors consider the various types of corrosion relative to the surrounding medium (air, water, soil, foodstuffs), and the resistance which the different covering materials offer to corrosive attack. They then proceed to consider the metallic coatings (zinc, aluminium, tin, lead, cadmium and nickel) in further detail. Finally, metal spraying is dealt with.

The Evolution of the Plating Bath. H. J. T. Ellingham. (Electrodepositors' Technical Society, May 15, 1935: Metal

Industry, 1935, vol. 46, May 17, pp. 539–540). The author traces the history of electro-deposition, from the introduction of the single-cell process in the early nineteenth century up to the present time. He includes references to the use of external batteries as the source of electrical energy, to the dynamo, and, more particularly, to the development of the plating solution in the direction of cyanide and sulphate baths.

Specifications for Cathodic Coatings. C. F. J. Francis-Carter. (Electrodepositors' Technical Society, Mar. 27, 1935: Metal Industry, vol. 46, June 7, pp. 629–631). The author divides specifications for plated coatings into two types: Process specifications, and inspection specifications. In the former the processes used in deposition are laid down, whilst in the latter full choice of manufacturing methods is allowed, but certain inspection tests are imposed. A third type, less rigid than the first, but more definite than the second, is also suggested. The author considers that specifications involving inspection only are to be preferred, but remarks that the tests should be properly defined. He then goes on to discuss the appearance, adhesion, thickness and porosity of cathodic coatings such as those of nickel and chromium.

Specifications for Anodic Deposits. S. Wernick. (Electrodepositors' Technical Society, Mar. 27, 1935: Metal Industry, 1935, vol. 46, June 7, pp. 631–632). Specifications for zinc and cadmium coatings are discussed. The author divides them into two classes: Process specifications, connoting rigorous control of the methods of deposition; and requirements specifications, which are imposed by the purchaser. The importance of the plating procedure is stressed, and the properties which are likely to be specified by the purchaser are stated. They are: Appearance, thickness or weight of the deposit, uniformity, adhesion, and protective value.

Protective Film on Ferrous Alloys. F. Fenwick. (Industrial and Engineering Chemistry, 1935, vol. 27, Sept., pp. 1095–1099). An electrometric test is described in which a cell is set up, one electrode of which is the metal under examination, the other being a suitable standard electrode. A chloride solution is slowly added to the solution in which the metal is immersed. The sudden change in potential which is observed marks an abrupt change, not in the specific ion concentration of the electrolyte, but in the character of the surface of the electrode. The method was applied to plain carbon, copper-bearing and stainless steels, the results being in general agreement with the known behaviour of these materials under conditions of atmospheric corrosion; further work on the method is, however, required.

Electrolytic Deposition of Copper-Nickel-Iron Alloys. H. Paweck, J. Bauer and J. Dienbauer. (Zeitschrift für Elektro-

chemie, 1934, vol. 40, pp. 857–862). The authors have determined the conditions necessary for the deposition of copper-nickel-iron deposits of any desired composition, including that corresponding to monel metal.

The Electrodeposition of Tin and its Alloys. Part I.—The Effect of Addition Agents on the Deposition of Tin from Alkaline Stannate Baths. L. E. Stout and A. Erspamer. (Electrochemical Society, 1935, Preprint No. 68–4), The main purpose of the research was to find the most suitable addition agent for tin plating from alkaline stannate baths. Lactic acid appears to give the best results, a concentration of 0·6 grm. per litre being suitable. Other addition agents which work quite well are malic acid, ethylene glycol and tartaric acid. The effect of sodium perborate, sodium biborate, casein, glucose, aloin, starch, triethanolamine and glycerine was also studied. Increasing the alkalinity of the bath reduces the weight of the deposit, and varying the concentration of the addition agent also has an effect on the weight. With some agents a certain concentration of the addition agent gives rise to a maximum weight of deposit, while with others a deposit of minimum weight forms.

The Electrodeposition of Chromium. E. A. Ollard. (Metal Industry, 1935, vol. 46, May 17, pp. 541–542 ; vol. 47, July 26, pp. 89–91 ; Aug. 23, pp. 185–188). The theoretical considerations underlying the electrodeposition of chromium are briefly dealt with. Reference is made to the chromium complexes present, the effect of sulphate additions, and the mechanism of electrolysis. The presence of sulphate and fluoride ions is known to have a strong influence on the process of electrodeposition. The author advances an explanation to account for the effect of the former, on the basis of the formation of colloidal chromium complexes round the cathode. The preparation and operation of the plating solution is subsequently dealt with, and details covering the type of anodes, suspension of the work, design of jigs, auxiliary anodes, preliminary treatment of the work, the control of the solution, the estimation of conductivity and control of the sulphates are furnished.

Mechanism of Chromium Deposition from the Chromic Acid Bath. C. Kasper. (Journal of Research of the National Bureau of Standards, 1935, vol. 14, June, pp. 693–709). In this paper evidence is presented which supports the view that the deposition of chromium occurs directly from the hexavalent state. Thus, relatively large additions of compounds of metals such as zinc, nickel, cadmium, iron, and copper have no such effect on the behaviour of the chromic acid bath as would be expected if deposition occurred from the trivalent state. The low efficiency and high polarisation in chromium deposition are accounted for, and the possible directions for improvement are indicated.

Theory and Practice of Chromium Plating. N. D. Birükoff, S. P. Makariewa and A. A. Timochin. (Korrosion und Metallschutz, 1935, vol. 11, Aug. pp. 172–179; Sept. 193–201). In the first communication the authors record their experimental investigation of the oxidation and reduction processes that take place during the deposition of chromium from chromic acid solutions containing additions of sulphuric acid. In the second communication, they give the results of their examination of the influence of interruptions during the electrolysis on the percentage yield of metallic chromium and on the potential across the terminals, and the periodic variations in the rate of evolution of hydrogen.

Nickel-Chromium Plating Automobile Parts. M. Yamaguchi. (Japan Nickel Review, 1935, vol. 3, July, pp. 405–420). The general procedure followed in representative nickel and chromium plating plants in America, England and on the Continent, is described. The information relates chiefly to nickel plating practice; the operating conditions employed in Europe and America are summarised.

Cadmium as a Means of Corrosion Protection. L. Nernst. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, July 27, pp. 920–922). The properties of cadmium, methods of using it, and the results of corrosion tests on it are discussed on the basis of the literature of the subject. The electrolytic production of zinc in Germany seems likely to provide large quantities of this metal as a by-product.

The Testing of Electrodeposited Coatings. L. C. Bannister. (Electrodepositors' Technical Society, Mar. 27, 1935: Metal Industry, vol. 46, June 28, pp. 701–703). A selection of representative methods of examining electrodeposits is outlined. The tests include examination for appearance, composition and thickness, and for physical properties such as adherence, hardness and corrosion resistance. In conclusion, it is suggested that the tests applied to electrodeposits should depend on the service which the article is required to give, and that the tests should, if possible, be reproducible.

The Measurement of the Thickness of Corrosion-Protective Coatings. E. Reitler. (Röhrenindustrie, 1935, vol. 28, July 11, pp. 73–74). (*See* E. Schweitzer and S. Kiesskalt, Journ. I. and S.I., 1934, No. I. p. 540).

Zinc Ammonium Chloride. Its Place in Modern Galvanising. H. G. Hobbs. (Iron Age, 1935, vol. 135, May 23, pp. 10–13, 84–90; June 13, pp. 22–27, 90; July 18, pp. 18–21, 92, 94). The early history of galvanising is traced, and the first records of the use of fluxes—such as hydrochloric acid and sal ammoniac—are given. Certain chemical considerations relative to the fluxing action are discussed. In the concluding article

in this series the author surveys the merits and demerits of zinc ammonium chloride as a flux in the galvanising of pipes, wire, and hollow-ware. He contends that zinc ammonium chloride is superior to other fluxes in the coating of pipes and of nails, but that there is no net gain when galvanising wire. Pails, buckets, drums and other fabricated articles may be successfully fluxed with zinc ammonium chloride if sufficient attention is paid to the proper degreasing, pickling, rinsing and draining of these products.

Fluxes for Galvanising. W. G. Imhoff. (American Metal Market: Sheet Metal Industries, 1935, vol. 9, Aug., p. 513). Hot-dip galvanising may be carried out with a molten flux all over the bath, a molten flux over part of the bath, or no molten flux on the bath at all. The conditions under which each type of flux covering is used are indicated.

Submersion Time *versus* **Quality of Hot-Dip Zinc Coatings.** W. G. Imhoff. (Iron Age, 1935, vol. 135, June 6, pp. 16–17, 124, 126, 128). In the process of galvanising, a large number of variables play a part, of which the period of immersion in the molten zinc is one. If all the other factors are kept constant the immersion time is found to influence the following: The spangle effect; the finish of non-spangled coatings; the thickness of the coating; the iron content of the coating; its weight, smoothness, lustre, and adherence; and the bending qualities of the product. Prolonged contact with the bath of molten zinc tends to destroy the spangle effect, and ultimately to roughen the surface; the coating also becomes thicker. A long immersion period increases the iron content of the zinc coating and makes it brittle; the longer the immersion time the poorer the lustre and adherence of the zinc coating. On the other hand, short immersion times produce a coating which is well spangled, smooth, thin, lustrous, highly adherent and free from excessive iron and brittleness.

Residuum Losses in Galvanising. W. H. Spowers, jun. (Industrial Heating, 1935, vol. 2, May, pp. 239–243). The sources of dross losses in galvanising are indicated. These losses are shown to be mainly due to improper containers, active fluxes, improper firing and oxide losses. It is suggested that in most cases high residuum losses could be greatly reduced.

Detection of Porosity in Zinc Coatings on Steel. G. Garre. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Aug., pp. 91–94). After touching on various methods of detecting porosity in zinc coatings on steel, the author gives details of a development of the electrolytic method. In this, the specimen is made the anode in an electrolytic cell containing an aqueous solution of potassium ferrocyanide (not ferricyanide) to which is added some magnesium sulphate to improve the conductivity:

Potassium ferrocyanide	20 grm.
Magnesium sulphate	1 grm.
Water	500 grm.

This solution attacks the zinc less than any other. Porosity is revealed by a blue coloration or deposit. Current is supplied by a 4-v. accumulator. Gas is evolved from both the anode and cathode (platinum for preference) as soon as current is passed if porosity or exposed iron exists, but the evolution on both electrodes stops as soon as the iron is completely coated with the blue deposit.

Hot-Dipped Tin Coatings. E. J. Daniels. (Transactions of the Faraday Society, 1935 : Sheet Metal Industries, 1935, vol. 9, July, pp. 425-428). The author suggests that the nature of the interfacial alloys may affect the formation and structure of hot-dipped tin coatings, and discusses the CuSn type of coating known to exist on tinned copper. Fluxing and the process of solidification are next dealt with, and attention is given to rippling and spangling.

Factors Influencing the Formation and Structure of Hot-Dipped Tin Coatings. E. J. Daniels. (Technical Publications of the International Tin Research Development Council, 1935, Series A, No. 17). Existing theories concerning the factors involved in the formation of hot-dipped coatings are reviewed, with particular reference to surface tension as a prime cause of wetting and adhesion. It is considered that the tendency to form a solid solution or an intermetallic compound is an operative factor. Attention is directed to the interfacial alloy which is produced and the effect of alloy additions to the tin. The fluxing process is discussed and it is pointed out that the flux should be molten at the working temperature. Rippling, and the general quality of the tin coating, are found to be related to the composition of the tin bath and the temperature of dipping.

Improvement in the Quality of Tinplate by Superimposed Electrodeposition of Tin. A. W. Hothersall and W. N. Bradshaw. (Chemistry and Industry, 1935, vol. 54, Sept. 20, pp. 320T-326T). An examination of the effect of electrodeposited tin coatings on the porosity of tinplate revealed that thin electrodeposited coatings diminish and thicker coatings eliminate porosity, provided the process is carried out in an alkaline bath. The following conditions were found to favour maximum reduction of porosity: (1) Efficient cleaning prior to plating; (2) a low current density during deposition. The plating bath was a sodium stannate solution containing sodium hydroxide.

Black Spots on Tin and Tinned Ware. S. Brennert. (Technical Publications of the International Tin Research and Development Council, 1935, Series D, No. 2). The so-called "black spots" sometimes noted on tin, and especially on tinned material, are due to an electrochemical corrosion process. On a free tin surface the formation of black spots does not begin until a certain potential against the solution (the "dissolution potential") is reached,

the value of which varies with the composition of the solution. The black spots can continue to grow, however, especially in very dilute solutions, at potentials less noble than the dissolution potential. The single potential of a freshly polished tin surface and that of a newly tinned surface generally represent values less noble than the dissolution potential; but, through the further growth of the air-formed film on the surface of the tin in neutral or nearly neutral solutions (*e.g.* sodium chloride solutions and milk), the potential may rise above the dissolution potential. At " weak points " of this film the tin can then corrode and black spots are produced. The black spots consist mainly of stannous oxide. The formation of black spots can be prevented by keeping the tin in contact with a less noble metal (*e.g.* zinc or aluminium), thus ensuring that the potential is kept below the dissolution potential.

The Use of Sodium Sulphite as an Addition to Alkaline Detergents for Tinned Ware. R. Kerr. (Chemistry and Industry, 1935, vol. 54, July 5, pp. 217T–221T). Experiments carried out on the corrosion of tin and of tinned copper in hot sodium carbonate and sodium hydroxide solutions revealed that the severity of the attack depends largely on the concentration of dissolved oxygen. The effect of various reducing agents was investigated, and, of those tested, sodium sulphite was found to be the most promising. If it is added to the alkaline cleaning solution the rate of attack may be reduced to nearly one-tenth. A ratio of 1 part of sodium sulphite to 10 parts of sodium carbonate or 4 parts of caustic soda is recommended. The sulphite appears to have the effect of removing the dissolved oxygen from the solution. The oxidation product, sodium sulphate, is harmless, while the sulphite has the advantages of being inexpensive, relatively stable in the solid state, and readily soluble in water or in alkaline solutions.

Examination of the Surface of Tinplate by an Optical Method. W. E. Hoare and B. Chalmers. (Iron and Steel Institute, Sept., 1935: this Journal, p. 135).

Coating Strip with Copper Lead for Ford Rod Bearings. (Iron Age, 1935, vol. 136, Aug. 8, pp. 26, 47). The Ford Motor Co., uses copper-lead for the connecting rod bearings of Ford motor car engines. Steel strip is passed down through a bath of molten metal having the following analysis: Lead 30 to 35, copper 63 to 68, iron 0·50 (max.), nickel 1 to 1·5 per cent. It emerges at the base of the coating furnace through a water-cooled graphite die which causes the steel strip to be uniformly coated with the copper-lead mixture. It is then cut to length and formed into semi-circles for the bearings.

Ford Granodizes Lamp Shells as Protection against Rust. B. Finney. (Iron Age, 1935, vol. 136, July 11, pp. 33–34). In the

electro-granodizing process a protective film, 0·0005 in. thick, of crystalline zinc phosphate, is deposited on the lamp shells. The latter are placed in the plating bath fed by a " granodizer " from an a.c. supply at a current density of 35–40 amp. per sq. ft. The time of immersion is 4½ min., at a temperature of 150° F. (The composition of the granodizing bath is not given.) After the granodizing treatment the shells are ready for enamelling. Granodizing is stated to render the articles highly rust-resistant.

A New Method for Improving the Surface of Castings. A. Karsten. (Giesserei, 1935, vol. 22, Aug. 16, pp. 411–412). After indicating the advantages and weaknesses of layer welding as a means of giving wear-, heat- or acid-resisting surfaces to cheap materials, the author refers to a new electric metal-spraying pistol, by means of which special alloy steels can be sprayed by the Schoop process.

The Hardness of Sprayed Metal Coatings. Kessner and T. Everts. (Zeitschrift für Metallkunde, 1935, vol. 27, May, pp. 104–107). The authors describe tests made with carbon steel, *V2A* steel and non-ferrous wires to determine the influence of various factors in spray-coating these metals by the Schoop process. Increasing the temperature of the basis metal lowers the Brinell hardness of a steel coating ; above 200° no further change occurs ; the structure of the metal is extremely spongy. Increasing the distance between the " pistol " and the work raises the scratch hardness of a steel coating, owing to the increased oxide content ; the hardness and oxide content increase with neutral and oxidising flames and high air pressure in the pistol. The particles of *V2A* steel scarcely oxidise during their flight, but with a longer path of travel they are less violently quenched and produce a softer layer. If the pressure of the gas in the pistol is raised the particles become hotter ; they are then quenched from a higher temperature, and so become harder. Further, as the particles on impact are more plastic and bind together better, the layer is less porous, and this influences the Brinell hardness. For steel the scratch hardness is greatest with high air pressures ; much oxide forms in spite of a reducing flame. A reducing flame always produces a beautiful evenly rough and lustrous coating, without visible oxide. At high air pressures the oxide content rises, owing to the turbulence of the stream ; the particles are more violently quenched and are hammered on to one another more powerfully, so that the porosity is lessened. For these reasons the scratch and Brinell hardnesses increase. Slow propulsion of the wire in the pistol increases the oxide content, as the wire melts deep down in the nozzle and the particles remain longer in the flame ; in steel coatings so produced, oxide nests may even form. The thick oxide coating produced under this condition heat-insulates the particles and prevents them from quenching, so that, despite the high oxide content, the

hardness of the layer decreases. The time interval allowed to elapse between the application of successive layers also affects the hardness. In conclusion, a brief tabular statement indicates the optimum conditions of spraying for the production of the hardest possible layers of steel, *V2A* steel and some non-ferrous metals.

A Study of Furnace Atmospheres and Temperature Gradients and Their Effect on Porcelain Enamelling. H. E. Ebright, G. H. McIntyre and J. T. Irwin. (Journal of the American Ceramic Society, 1935, vol. 18, Oct. pp. 297–302).

Cast Iron for Vitreous Enamelling. J. W. Gardom. (Institute of Vitreous Enamellers: Foundry Trade Journal, 1935, vol. 52, July 13, pp. 399). The author considers some of the factors upon which depends the suitability of cast iron used for vitreous enamelling. He draws the following conclusions: (1) That the surface should be smooth and free from blemishes; (2) that the casting should be sound; (3) that occluded gases should be removed; (4) that the structure should be stable, *i.e.*, composed of graphite and ferrite produced by annealing an iron of suitable composition; (5) that the silicon content should be sufficiently high to avoid the possibility of graphite entering into solution in the iron at the enamelling temperature; and (6) that the graphite should be in as fine a state of division as possible.

Common Defects in the Vitreous Enamelling of Cast Iron, Sheet Iron and C.R.C.A. Steel. C. P. Stone. (Institute of Vitreous Enamellers, Sept. 27–29, 1935; Foundry Trade Journal, 1935, vol. 53, Oct. 3, pp. 253–255). Attention is given to the preparation of the sheets, the ground coat dipping, first firing, enamel spraying, firing and inspection, and to the same processes when applied to the enamelling of cast iron by the wet method.

A Study of Certain Factors Affecting the Proper Firing of Porcelain Enamel on Steel or Iron. H. E. Ebright and G. H. McIntyre. (Industrial Heating, 1935, vol. 2, June, pp. 293–298). The results of a series of tests extending over a period of several years are presented. A gas sampling apparatus for obtaining the composition of the air from the furnace muffle was used, and the figures obtained under a variety of weather conditions, and of furnace operation, are reproduced. These results are examined, and it is shown that atmospheric and thermal conditions have a marked influence on the quality of the enamel produced. Enamel defects such as scumming, blistering and particularly copperheading may be frequently attributed to contaminated atmospheres or incorrect temperature conditions.

The Use of a Nickel Dip in Enamelling Practice. J. E. Hansen and J. T. Irwin. (Journal of the American Ceramic Society, 1935, vol. 18, Aug., pp. 225–229). The compositions of various nickel salt solutions now being used in enamelling practice are given, with reasons for their use. The advantages to be gained

by employing these solutions are : Better adherence of enamel to the metal, decreased fishscaling, and control of copper-heading conditions. Methods for the control of the *p*H value and for the analysis of solutions are furnished.

Mechanics of Enamel Adherence. XI.—Application of the Theory of Dendritic Adherence to the Development of White Ground Coats for Sheet Steel. W. L. Housley and R. M. King. (Journal of the American Ceramic Society, 1935, vol. 18, Oct., pp. 319–320).

Investigation of Some Physical Factors Affecting the Draining Qualities of Ground-Coat Enamels. W. N. Noble and G. S. Lindsey. (Journal of the American Ceramic Society, 1935, vol. 18, Oct., pp. 308–314).

The Fineness Distribution of Vitreous Enamel as Affected by Variations in Grinding. E. P. Poste. (Journal of the American Ceramic Society, 1935, vol. 18, Oct., pp. 303–305).

Opacity Development in Cover Enamels for Sheet Steel. J. D. Tetrick. (Journal of the American Ceramic Society, 1935, vol. 18, Oct., pp. 314–318). The results of a study of opacity development in cover enamels for sheet steel are given, and the effects of firing time, firing temperature, gauge of base metal and of some opacifiers are discussed.

Antimony Compounds as Substitutes for Tin Oxide as the White Colouring Agent of Enamel, and the Examination of Antimony-Bearing Enamels from the Hygienic Point of View. K. Beck. (Korrosion und Metallschutz, 1935, vol. 11, June, pp. 132–133).

Note on the Relation between Cobalt and Nickel-Oxide Content and Reboiling of Sheet-Steel Ground Coats. J. M. Cayford and R. M. King. (Journal of the American Ceramic Society, 1935, vol. 18, July, p. 224). Varying amounts of cobalt and nickel oxides were added to a selected enamel, and the duration and persistence of reboiling were determined. The oxides were found to exert a definite influence on the reboiling phenomenon.

Copperheads or Iron Oxide Defects in Porcelain Enamel. J. J. Canfield. (Iron Age, 1935, vol. 136, Aug. 15, pp. 30–34). Copperheads in porcelain enamel may be attributed to the presence of local spots of iron oxide. A section of an enamelled plate in the underfired state reveals the presence of a layer of undissolved iron oxide at the junction between the enamel and the metal. As the firing proceeds the enamel fuses and the iron oxide layer is absorbed by the enamel. During this stage bubbles form, increasing in size with the duration of firing. Experiments carried out on melts of enamel containing iron oxide show that excess iron oxide tends to separate in the crystalline form. These crystals have been identified in copperhead structures, and are frequently, though not always, associated with copperheading.

This defect may be due to a variety of causes, such as inadequate cleaning, a thin enamel coat, improper drying or heating rates, and so on.

Asphalt Pipe Coating Tanks Heated by Tubular Electric Immersion Units. (Steel, 1935, vol. 97, Aug. 5, pp. 52–53). For the purpose of satisfactorily coating large pipes with asphalt a tubular immersion heater has been devised. The heater is in three sections, thus enabling independent temperature control to be maintained. The tank containing the asphalt is insulated to reduce radiation losses.

Some Notes on Heat-Resisting Rust-Preventing Paints. F. J. Peters. (Korrosion und Metallschutz, 1935, vol. 11, Aug., pp. 179–183). The author first discusses the difficulty of finding suitable media for paints for use at temperatures up to, say, 400° C.; heat resistance and rust protection demand two contradictory properties, namely, thinness and oil-richness. He then touches on the colouring matters employed, and goes on to deal with the behaviour of paints such as aluminium-bronze and zinc-dust paints. He points out that heat-resistant paints should never be applied by spraying.

PHYSICAL AND CHEMICAL PROPERTIES.

PROPERTIES AND TESTS OF CAST IRON.

A New Machine for Testing Cast Iron. (Iron and Steel Industry, 1935, vol. 8, Sept., pp. 462–463). The machine is designed to deal with test bars comparable in diameter to the thickness of the castings under test. Bars up to 10 mm. in diam. may be tested under a bending load, the stress being applied by a shot-weighted lever. The machine is suitable for testing light castings (such as stove, household and sanitary goods) in which the wall thickness of the article is approximately the same as that of the specimen—*i.e.*, between 2·5 and 10 mm. in diameter.

Test-Bars. R. S. MacPherran. (Transactions of the American Foundrymen's Association, 1934, vol. 42, pp. 476—484). A brief review of the present state of test-bars for the testing of cast iron.

Relationships in Cast Iron Test Results. G. L. Harbach. (Institute of British Foundrymen, July 2 to 5, 1935: Foundry Trade Journal, 1935, vol. 53, July 25, pp. 61–63, 72; Aug. 1, pp. 85–87). It is pointed out that the most common test applied to cast iron is the transverse test. However, for engineering purposes it is desirable to estimate the tensile strength, and this may be done if a relation between the transverse and tensile strengths is established. The factors which determine this relation are discussed; they include the size, surface finish (whether machined or as-cast) and grade of iron. In the light of experiments carried out by the author, it is concluded that the ratio between the transverse modulus of rupture and the tensile strength falls with increasing strength, from 2·3 at 10 tons per sq. in. tensile strength to 1·6 or less at over 20 tons tensile strength. The ratio is not constant within closer limits than ± 8 per cent., except when the iron is produced under closely controlled conditions. Machined bars give higher ratios than bars in the as-cast condition. In the case of test bars cast on end no great difference in the strengths of the top and bottom of the bars was found. Variation in span appears to have a greater effect, the average results of transverse tests being highest when the span is shortest. Tests on bars with different cross-sections revealed that 1-in. × 1-in. bars were stronger than 2-in. × 1-in. bars, while bars of 1·2 in. diam. were also superior in this respect. In order to determine the influence of sectional area, test bars of various diameters were cast and examined. The results show

that although increasing section leads to loss in strength, there is less relative loss with high strength irons; for example, the material at the centre of a thick bar was found to retain 86 per cent. of the tensile strength of the original bar. The author's statements are supported by curves and tables relating to the investigations which were carried out.

Grey Cast Iron. J. W. Bolton. (Foundry, 1935, vol. 63, Apr., pp. 18–21, 48, 50, 53; May, pp. 27–30, 66; June, pp. 22–23, 78; July, pp. 21–23, 48, 50; Aug., pp. 38–39, 80; Sept., pp. 32–33, 84, 86). The metallurgical and metallographical characteristics of grey iron and of the constitutents contained therein are described in a general way, and illustrated by micrographs.

Yielding of Cast Iron as seen during the Bending Test. S. Ikeda and K. Okamura. (Transactions of the Society of Mechanical Engineers, Japan, 1935, vol. 1, May, pp. s27–s28). The authors analyse the results obtained by subjecting cast iron to a bend test, during which the fibre strains were measured. They show that yielding comparable to that exhibited by mild steels occurs.

The Notch-Sensitivity of Cast Iron under Impact Stressing. E. von Rajakovics. (Giesserei, 1935, vol. 22, Sept. 13, pp. 458–459). The author discusses the lack of agreement in the results of various investigators' determinations of the influence of a notch on the specific energy of impact of specimens of cast iron, and describes his own experiments made in an attempt to elucidate the matter. He concludes that while cast iron appears in general to be insensitive to notch action under tensile alternating stresses, yet under impact stresses it has a not inconsiderable notch-sensitivity.

Impact Resistance and Other Physical Properties of Alloy Grey Cast Irons. G. P. Phillips. (American Foundrymen's Association, Aug., 1935, Preprint No. 1). To determine the impact resistance of alloy cast irons, samples from 38 cupola heats and 20 electric furnace heats were tested. The alloy additions used were chromium, molybdenum, nickel, copper and titanium. Ten heats of the electric furnace irons were heat-treated. The effects of the alloying elements on the physical properties are compared and discussed. The results indicate that, in the case of irons cast grey, and omitting austenitic iron, fairly high-carbon nickel-molybdenum irons that may contain some chromium have good impact-resisting properties. The highest impact resistance was obtained on irons that were cast white and converted to grey irons by annealing. Austenitic irons of low hardness have high impact resistance. Transverse resilience correlates fairly closely with impact resistance, and apparently the single-blow impact test made on machined bars correlates with the transverse

resilience more closely than does the repeated drop test on unmachined bars.

The Work of Rupture and the Mechanical Strength of Grey Cast Iron. J. N. Alcacer. (Transactions of the American Foundrymen's Association, 1934, vol. 42, pp. 655–671).

The Influence of Prestressing on the Strength Properties of Cast Iron. A. Leon and A. Slattenschek. (Giesserei, 1935, vol. 22, July 19, pp. 353–357). The authors have determined the influence of a previous application of stress (tension, compression or torsion) on the mechanical properties of cast iron. A compressive stress approaching the compressive strength, applied in the direction of pulling, decreased the tensile strength obtained by about 9 per cent.; the elastic modulus in tension, related to the total deformation, decreases by about 38 per cent. In the compression test the volume of the specimen increases; prestressing in compression or in tension causes a reduction of this expansion, and in the case of the iron tested the effects were equal. When specimens were prestressed in compression followed by tension, the resulting decrease of the expansion in the subsequent compression test was equal to the sum of the decreases caused by compression alone and by tension alone. A compressive prestress augments the effect of tensile prestressing on the density of cast iron. Prestressing had not much effect on the compressive strength. Storage of the specimens between prestressing and subsequent testing reduced considerably the influence of tensile or torsional prestressing on the compressive strength. The angle of fracture in the compression test was not noticeably altered by compressive or tensile prestressing, and compressive prestressing was also without influence on the angle of fracture in the tensile test. Hence the ratio of the true compressive strength to the true tensile strength is not practically affected by either compressive or tensile prestressing.

The Mechanical Properties of Cast Iron at High Temperatures. R. Bertschinger and E. Piwowarsky. (Giesserei, 1935, vol. 22, July 5, pp. 325–333). After reviewing the literature of the subject, the authors give an account of their experiments, in which the bend strength, compressive strength and Brinell hardness at raised temperatures were determined on a number of approximately eutectic grey cast irons with varying silicon and high carbon contents (and relatively unfavourable graphite development) and the smallest possible amounts of phosphorus, sulphur and manganese. The silicon contents varied from 0·5 to 3 per cent., the casting temperature was 1,350° C., the mould temperature was 150°, and the testing temperature ranged from 20° to 560° C. The occurrence of a minimum in the strength values between room temperature and 400° was not connected with the casting stresses. In the rapid tests described, a sharp

fall in the strength set in between 380° and 420°. In the range examined, no specific influence of the silicon on the hot strength could be observed. Coarse graphite apparently depreciated the hot strength; the finer the graphite, the higher were the absolute strength values, even at 500°. At 550° the bend strength and compressive strength fell to half their values at room temperature. In the temperature range 20° to 350°, the elastic modulus behaved erratically; in tests on one part of the material it fell steadily, while in the other part it developed a maximum at 100°–150°. Coarse graphite reduces the elastic modulus to values of 3,000–4,000 kg. per sq. mm.

Austenitic Grain-Size in Cast Iron. D. W. Murphy and W. P. Wood. (American Society for Metals, Oct., 1935, Preprint No. 15). Evidence is presented which shows that the properties of grey cast iron are greatly affected by variations in the austenitic grain-size of the material. Increasing the austenitic grain-size improves the as-cast strength and hardness. It is shown that the hardenability of grey cast iron increases with increasing austenitic grain-size and that the effects of grain-size on the strength and hardness may possibly be due to the effect of grain-size on the hardenability. The importance of the question of hardenability in grey cast iron, particularly with the greater application of heat-treating, is discussed.

Thermal Expansion Characteristics of Some Nickel Cast Irons. T. J. Wood. (Transactions of the American Society for Metals, 1935, vol. 23, June, pp. 455–468). The author has investigated the thermal expansion of a series of cast irons ranging in nickel content from 0 to 70 per cent. In the ferritic cast irons, limitations in the nature of less resistance to heat and corrosion narrow the applicability of these irons to fairly low temperature ranges and to mildly corrosive conditions. One exception in this group is the 4·5 per cent nickel, 1·5 per cent chromium white cast iron, which has remarkable heat-resistant properties. The non-magnetic austenitic cast irons have high expansion characteristics, and in order to obtain equivalent resistance to heat and corrosion with lower expansivities the nickel content should be within the range 27 to 45 per cent. The lowest expansion rates of any of the irons investigated were obtained in the range 34–35 per cent. nickel. The substitution of nickel plus copper (in the Monel ratio) for an equivalent amount of nickel in cast iron produces expansion rates comparable to those of the straight nickel alloys with 15 and 20 per cent., but at higher ranges the expansivities are of a higher order than those of the corresponding copper-free cast irons.

The Wear Resistance of White Cast Iron. O. W. Ellis, J. R. Gordon and G. S. Farnham. (American Foundrymen's Association, Aug., 1935, Preprint No. 4). Tests were carried out in a

ball mill on white cast iron balls of 1-in. diameter, using silicon carbide as the grinding medium. The results show that the higher the carbon content or, more generally, the lower the primaustenoid content of white cast iron, the less consistent are the results of the wear tests. The higher the carbon content of white cast iron, other things being equal, the lower is its wear resistance under the condition of the test described. Variations in the manganese content up to about 1 per cent. are practically without effect upon the wear resistance, and no benefits accrue from adding up to 6 per cent. of manganese to white cast iron containing ledeburite. An increase of silicon from about 0·75 per cent. to about 1·5 per cent. has little or no effect on the wear resistance of white cast iron but results in a remarkable increase in that of mottled iron. It is suggested that this increase in wear resistance is due to the solution in austenite, at the stable eutectic temperature, of silicon which is made available as a result of the dissociation of eutectic iron carbide and the consequent formation of graphite, but which would have been retained in solid solution in the eutectic iron carbide if metastability were established. Sulphur was found to be deleterious in its effect on the wear resistance of white cast iron.

A Study of the Machining Problem with Particular Reference to Cast Iron. H. H. Beeny. (International Foundry Congress, Brussels, 1935 : Foundry Trade Journal, 1935, vol. 53, Sept. 26, pp. 227–231). The author discusses the metallurgical aspects of machinability, particularly the effects of chemical composition, physical structure, static strength, Brinell hardness, ductility, dynamic strength (Izod value), work-hardening and frictional properties, and proceeds to consider the case of grey cast iron at greater length ; the influence of carbon, sulphur, manganese, silicon, and phosphorus in particular receives attention. It is concluded that the low tensile strength, impact resistance and ductility of cast-iron are largely responsible for its good machinability ; that free (graphitic) carbon facilitates machining, whilst combined carbon renders it more difficult ; that phosphorus improves machinability ; and that for maximum machining quality a low-carbon, high-silicon iron in the annealed state is the most suitable.

Superheating Cupola Melted Iron. G. P. Phillips. (Foundry, 1935, vol. 63, Aug., pp. 33–35, 72, 74). The author presents the results of an investigation carried out to determine the effect of superheating on the properties of cast iron. Three series of tests were made on the following cupola-melted irons : First series, nickel-chromium alloy—the cupola mixture contained 20 per cent. of steel, 37 per cent. of pig iron and 43 per cent. of scrap iron ; second series, containing a small amount of incidental nickel and chromium—the cupola mixture contained 53 per cent.

of pig iron and 47 per cent. of scrap iron ; third series, nickel-chromium alloy—cupola mixture same as first series. The superheating was carried out in an indirect-arc, rocking electric furnace. The results of the tests indicate that the superheating of cupola-melted, steel-mix, alloy iron, melted at high temperatures in the cupola, gives no particular advantage except that of higher temperature only. In the case of plain iron or very low alloy content iron, made with a pig-scrap mixture in the cupola, a decided improvement was obtained by superheating the metal. Microscopic examination of the various metals showed little change in graphite size and distribution in the alloy irons, but did show improvement in the plain pig-scrap iron of the second series.

Factors Affecting the Structure and Properties of Grey Cast Iron. A. Di Giulio and A. E. White. (American Foundrymen's Association, Aug., 1923, Preprint No. 9). The authors have investigated the effects of superheating and pouring temperature on the structure and properties of grey cast iron. The improved physical properties obtained through superheating and by pouring at the correct temperature are ascribed to a refinement of the graphite flakes.

Chilling Properties of Cast Iron. W. H. Spencer. (Transactions of the American Foundrymen's Association, 1934, vol. 42, pp. 508–524). The author has studied the chilling properties of cast irons using two sizes of test-bar. It was found that variations in chill depth due to changes in pouring temperature were very slight. Sand, used for ramming the chill test specimens, also seemed to have little or no effect on the depth of chill. The effect of composition on chill may be summarised as follows : Chill depth was increased by high sulphur or manganese, and by ladle additions of antimony, bismuth, brass, cadmium, cobalt, lead, tin and vanadium. It was decreased by high silicon or total carbon, or phosphorus contents. The ratios of silicon to total carbon and of manganese to sulphur were qualitatively indicated by the chill test.

Influence of the Accompanying and Alloying Elements Silicon, Phosphorus, Nickel, Chromium, Molybdenum, Tungsten and Copper on the Section Susceptibility of Grey Cast Iron. E. Hugo, E. Piwowarsky and H. Nipper. (Giesserei, 1935, vol. 22, Aug. 30, pp. 421–428 ; Sept. 12, pp. 452–458). The authors prepared a series of grey irons with various contents of the usual and also alloying elements and made castings in a box shape with the walls of different thickness ; specimens were cut from the different sections, their mechanical and other properties were determined, and the influence of the section thickness on the results obtained was evaluated. The representation of the absolute values of the strength properties, the hardness and the carbide content in

relation to the section thickness was carried out in the logarithmic manner suggested by Coyle, Jungbluth and Heller. The tests confirm the adverse influence of silicon and phosphorus and the favourable effect of nickel, particularly when it partly replaces the silicon. In the presence of chromium, nickel is particularly beneficial. In the absence of nickel the silicon, added to compensate the hardening action of chromium, must not be raised too much. Molybdenum has a similar effect to that of nickel, but is less effective in equalising the hardness. Tungsten and copper have a beneficial effect in regard to the influence of the section thickness, though it is slight. A new method of testing the shear strength is described, and its advantages are pointed out; the variation of this property within the thickness of the section is shown by curves. The results of the shear test confirm the influence of the elements revealed by the other tests. On comparison with the values given by the Sipp-Rudeloff and Frémont methods straight-line relations were obtained, demonstrating noteworthy agreement between the results obtained by the Sipp-Rudeloff and the present new method. Further, some relations between various properties of cast iron were calculated and critically evaluated. Finally, the results obtained with a few specially cast bars were compared with those given by the specimens from the box-shaped castings.

The Effect of Copper in Grey and Malleable Cast Iron. A. J. N. Smith. (Bulletin of the British Cast Iron Research Association: Foundry Trade Journal, 1935, vol. 52, June 13, pp. 402–403). This paper constitutes a review of the literature on the above subject.

The Influence of Aluminium on the Properties of Ordinary Cast Irons. M. Ployé. (Revue de Métallurgie, Mémoires, 1935, vol. 32, June, pp. 248–270; July, pp. 302–320). The author presents the results of a research to determine the influence of aluminium additions on the properties of cast iron. For quantities of from 0·02 to 0·25 per cent., added to cupola-melted metal, the fluidity increased and the chilling effect decreased; additions of 0·05 per cent. were without effect on the mechanical properties and microstructure, but larger amounts improved the properties slightly and caused a slight development of graphite lamellæ. The best way of adding the aluminium was to add it hot to the ladle, and stir well before casting. The effects of large additions (0·5 to 18 per cent.) of aluminium to ordinary cast irons melted in a crucible were very variable, according to the amount added and the analysis of the iron treated. In general, however, the properties all improved with up to 2 per cent. of aluminium; above that figure the only features of interest were the resistance to shear (up to 8 per cent.), the increase of hardness (up to 15 per cent.), and particularly the very marked increase in resistance

to corrosion in sea-water and salt spray and at high temperatures. With 18 per cent. of aluminium the alloys were magnetic. Unfortunately the high-aluminium alloys are very difficult to work, and the improvement in certain directions is nullified by the deterioration of the mechanical properties.

The Influence of Hydrogen on Cast Iron. W. Baukloh. (Giesserei, 1935, vol. 22, Aug. 16, pp. 406–409). After touching on the importance of the subject from the practical point of view, the author discusses the chemical attack of hydrogen on the carbon contained in cast iron, the decarburisation of white and grey cast iron, and the influence of alloying elements on the decarburisation by hydrogen; he also deals with the connection between decarburisation and diffusion. Lastly, he considers the hydrogen permeability of grey and alitised (aluminium-coated) cast iron, and compares it with that of steel, copper and nickel.

The Influence of Phosphorus on the Mechanical Properties of Grey Cast Iron at Raised Temperatures. M. Paschke and F. Bischof. (Giesserei, 1935, vol. 22, Sept. 13, pp. 447–452). After reviewing the literature of the subject the authors describe their experimental arrangements for testing the influence of phosphorus on the mechanical properties of cast iron at raised temperatures, and record the tensile strength, hardness and notch-toughness values obtained at temperatures up to 600° C. and over. The influence of the phosphorus content and of the silicon/manganese ratio on the shapes of the tensile-strength and hardness curves is shown; the absolute values of the silicon and manganese contents also influenced the tensile-strength curve on account of a displacement of the whole curve to higher values. Despite the constancy of the test conditions no relation between the phosphorus content and the shape of the notch-toughness curve at raised temperatures could be recognised.

Alloy Cast Irons. J. E. Hurst. (Metallurgia, 1935, vol. 12, May, pp. 15–18). Brief mention is made of early attempts to produce alloy irons, and the author goes on to enumerate the composition and properties of cast irons containing nickel, chromium and molybdenum additions, either singly or together. Alloy cast irons for hardening and tempering are next described, together with those suitable for nitrogen hardening. Finally, martensitic and austenitic cast irons, produced by higher alloy additions, are dealt with.

Iamzite Cast Irons. (Iron and Steel Industry, 1935, vol. 8, Aug., p. 425). Iamzite cast irons are a range of alloys produced in the foundry of Messrs. John Williams and Sons (Cardiff), Ltd., East Moors, Cardiff; they include alloys suitable for resistance to heat, wear and corrosion; cast irons for high duty and with non-magnetic properties are also supplied.

German Progress in Piston Rings, Brake Drums and other

Automobile Castings. W. A. Geisler. (Institute of British Foundrymen: Foundry Trade Journal, 1935, vol. 53, Aug. 8, pp. 97–101, 104). The author compares single-cast with centrifugally cast piston rings, and suggests that for high-duty alloy irons the centrifugal process should be used, the single casting method being suitable for general purposes. Of the alloying elements, chromium, vanadium and molybdenum appear to give the most satisfactory results; they also improve the heat-resisttance, or resistance to loss of tension on heating. Similarly, centrifugally and single-cast rings are found to be superior to ordinary bushed cast rings in this respect. The radial pressure curves for single-cast heat-treated piston brings are more uniform than those for bushed rings. These differences persist during service. Piston ring hysteresis is greatest for bushed cast rings and least for centrifugally cast rings.

Cast iron may equally well be employed for cylinder liners and brake drums; for the latter purpose the rigidity and wear resistance of cast iron are especially valuable. Many composite drums have been invented in the endeavour to retain the advantages of the cast-iron drum whilst utilising the low weight and higher strength of steel. The author describes several of these composite drums.

German Advances in Automobile Castings. W. A. Geisler. (Giesserei, 1935, vol. 22, Sept. 13, pp. 460–466). The author describes improvements that have been made in the quality of cast-iron piston rings, cylinder liners and brake drums for automobiles.

Cylinder Liners. T. R. Twigger. (Automobile Engineer, 1935, vol. 25, May, pp. 179–181). Some results of tests on cylinder liners of centrifugally cast iron, in the heat-treated and unhardened conditions, also austenitic cast-iron liners, are presented. The tests were carried out under actual service conditions, and the results indicate that austenitic cast-iron cylinder liners and piston rings give superior wear resistance to ordinary cast irons.

High Test Alloy Cast Iron for Machine Tools. F. J. Dost. (Metal Progress, 1935, vol. 28, Aug., pp. 47–49). An account of the production of high-test alloy cast iron for lathes and similar machine tools is given. The type of iron employed is a semi-steel having the following analysis: Total carbon 2·79, combined carbon 0·61, silicon 2·05, nickel 1·50, manganese 0·85, phosphorus under 0·20 per cent. The charge is a mixture of selected pig iron and rail scrap; nickel shot is added in the ladle in order to improve the quality of the iron.

Cast Iron in Machine Construction. J. D. Watson. (Machinery (Lond.), 1935, vol. 46, July 18, pp. 488–490). Despite the extended use of alloy steels, drop forgings, malleable castings and weldings, cast iron still continues to be the chief material used in machine

construction. The author shows that nickel cast irons are particularly suitable for machine parts which are required to be strong, resistant to corrosion or abrasion, and amenable to heat treatment. He deals with twelve types of alloy cast iron, all containing nickel, together with other alloying elements such as chromium and silicon.

Alloy Cast Irons Win Place in Mining. H. Strom. (Engineering and Mining Journal, 1935, vol. 136, June, pp. 290–294). The author shows that the use of alloying elements in cast iron has served to rehabilitate cast iron as an engineering material, and proceeds to cite the properties and applications of the more important alloy cast irons, with particular reference to mining practice.

Endurance Limit of Blackheart Malleable Iron. E. G. Mahin and J. W. Hamilton. (American Foundrymen's Association, Aug., 1935, Preprint No. 2). The endurance limit of blackheart malleable iron was found to be higher than that reported by previous investigators. The form and possible influence of graphite nodules are briefly discussed.

Some Transient Phase Changes during the Graphitising Reaction. H. A. Schwartz, H. H. Johnson and C. H. Junge. (American Society for Metals, Oct., 1935, Preprint No. 6). The authors have attempted to follow the changes in composition and specific volume of the cementite and solid-solution phases during the process of graphitisation in the light of their own metallographic and density investigations. Material partially graphitised at 925° C. is converted on cooling and holding below A_1 into graphite plus Fe_3C plus decomposed solid solution (granular pearlite) of approximately metastable eutectoid composition. At the elevated temperature, however, the cementite decreases progressively in carbon content and increases in density, the latter remaining a simple function of the former. The stable solid solution saturated at 925° C. is less dense than the saturated metastable solution by about 1 per cent.

PROPERTIES AND TESTS OF IRON AND STEEL.

Tests and Testing. C. A. Kellogg. (Wire and Wire Products, 1935, vol. 10, Sept., pp. 411–413). Some of the problems which arise in the testing of wire are commented on.

Compound Lever Testing Machine for Russia. (Engineer, 1935, vol. 160, Aug. 30, p. 226). A description of a 180,000 kg. testing machine by W. and T. Avery, Ltd., Birmingham, for export to Russia, is given. It is of the compound lever type capable of carrying out tests in tension, compression and trans-

verse bending. There are four straining rods, a hydraulic cylinder for imposing the test load, and a system of weighing levers for indicating the load applied. Tension specimens up to 6 m. stretched length can be accommodated. The strain is applied by a hydraulic cylinder and ram, the latter being connected to a crosshead which receives the four straining rods of the machine.

Mirror Extensometer, Martens-Hesse Type. (Génie Civil, 1935, vol. 107, Aug. 17, pp. 161–162). A brief description is given of the principles of the Martens-Hesse mirror extensometer.

New Spring Tester. (Iron and Coal Trades Review, 1935, vol. 131, Aug. 23, p. 278). This machine, known as the "Elasticometer," is built in two load ranges, namely, 0 to 500 lb. and 0 to 750 lb. Provision is made for testing both tension and compression springs, automatic load indication by means of a pendulum being adopted.

Strength due to Shape. A. Thum and W. Bautz. (Stahl und Eisen, 1935, vol. 55, Sept. 26, pp. 1025–1029). Earlier methods of testing materials made use of procedures which had nothing in common with the actual stress conditions existing in practical machine construction. Their aim was to determine characteristic properties of the material and to assess its quality and usefulness on this basis. More recently the individual regularity of the strength of various constructional shapes has been more and more recognised, and it has been determined that this has an influence often much greater than that of the material itself. Thus it is found that there are no strength properties conditioned solely by the material alone, but that both the material and the shape contribute to the strength. It is therefore necessary to determine, for various materials, the strength due to shape (*Gestaltfestigkeit*) under stresses resembling service conditions, and thence to give the constructional engineer a usable basis for his calculations. So long as data regarding the strength due to shape of various structural parts are inadequate, the influence of the shape must be allowed for approximately by the use of experimentally determined coefficients, particularly the notch sensitivity. The authors give examples to show the influence of the shape of test-pieces and structural parts on the behaviour of material under tensile and alternating stress; they discuss the effect of the prevention of flow by unequal stress distribution under tensile and alternating load, and give consideration to the nature of the coefficient of notch sensitivity.

Experiments at Pressures of 5,000 Kg. per sq. cm. I.—Fundamental Determinations of the Effective Cross-Section of Pressure Balances. II.—Absolute Determination of the Compressibility of Solid Bodies. H. Ebert. (Physikalische Zeitschrift, 1935, vol. 36, pp. 385–392). In the first part the author describes the determination of the effective cross-section, &c., of a pressure

balance for 5,000 kg. per sq. cm. with unpacked pistons, by means of a differential method. In the second part he gives the results of the determination, by a comparator method, of the compressibility of various single crystals, including iron, and polycrystalline materials, including steel. In general, the values for single crystals are slightly higher than those for the same material in the polycrystalline condition, but the variation of the compressibility with the pressure is less. The actual pressure was measured with a resistance manometer analogous to the resistance thermometer. The changes of length, in millimetres per metre, caused by the pressures applied are tabulated for all the materials tested.

The Tensility of Soft Steel. H. D. Hibbard. (Industrial Heating, 1935, vol. 2, Feb., pp. 73–76). The author reviews the results of numerous investigators on the tensile strength of plain low-carbon steels at temperatures from −80° to 1,500° C. The data illustrate the limitations of such steels for structural and mechanical applications at elevated temperatures.

A Study of the Elastic Properties of Certain Specimens of Mild Steel. J. E. Hurst. (Metallurgia, 1935, vol. 12, July, pp. 85–86). Gaps were cut in rings of mild steel subjected to quenching treatments, and the gap movements recorded. The rings were examined for internal stress, modulus of elasticity, permanent set, and hysteresis; the results so obtained (for quenching temperatures of 500° to 750° C.) are compared with those yielded by rings in the as-received condition.

Study of the Elastic Limit of Structural Steels. G. Delbart and H. Francelle. (Revue de l'Industrie Minérale, Mémoires, 1935, No. 355, Oct. 1, pp. 463–472). The authors present the results of their tests made on a number of plain and alloy structural steels after various heat treatments in order to investigate the relations existing between the apparent elastic limit and proof stresses of 0·1, 0·2 and 0·5 per cent., and, further, to compare various special steels in order to be able to meet the requirements demanded by the French War and Naval Departments. Heating followed by slow cooling, when it produced a granular ferrite-pearlite structure, led to confusion of the proof stresses and resulted in a load-extension diagram with a horizontal portion. Heating followed by air-cooling gave the same results if the structure remained granular; in the case of low-alloy steels which air-hardened slightly (an increase of about 10 kg. per sq. mm. above the annealed value) the diagrams had no horizontal part, the 0·1, 0·2 and 0·5 per cent. proof stresses were distinct from one another, and the structures were sorbitic. Martensitic quenching accentuated the difference between the 0·1 and 0·2 per cent. proof stresses. Tempering and air-cooling after air-quenching or martensitic quenching tended to diminish the difference between the 0·1 and 0·2 per cent. proof stresses,

and these limits tended to coincide when the temperature approached Ac_1; when the temperature exceeded Ac_1 these limits remained confused if the steel scarcely hardened and the structure remained granular (carbon and silico-manganese steels), but separated again if the steel hardened partially in air (nickel-chromium, chromium-molybdenum, nickel-chromium-molybdenum special steels).

It appeared that the granular structure was not the only intervening factor, because tempering after martensitic quenching tended to close up the 0·1 and 0·2 per cent. proof stresses and produce curves with less steep slopes. Also, a nickel-chromium-molybdenum steel after furnace-cooling gave a load-extension curve without a horizontal part. Nevertheless, the 0·1 and 0·2 per cent. proof stresses approached one another more closely as more and more free ferrite was present in the structure. It was noteworthy that the diagram for steel quenched in the martensitic condition never showed any tendency to flatten out, whereas those for steels in which the cementite was clearly separated from the solid solution tended to bend over to give, if not a horizontal portion—which shifted upwards and to the right as the structure became finer and more homogeneous—at least a clear-cut point of inflexion with a flattening of the curve. By slowing up the application of the load the various limits were brought closer together. For the tempered and annealed conditions the locus of the 0·1 per cent. proof stress appeared to be linear. The ratios of the 0·1 and 0·2 per cent. proof stresses to the tensile strengths were highest for steels with homogeneous sorbitic or osmonditic structures, and lowest for those with granular (ferrite-pearlite), that is, annealed, structures. The moduli of elasticity varied only between 20,000 and 25,000 kg. per sq. mm., being low for martensitic, and high for osmonditic structures.

The Effect of Magnetisation on Young's Modulus of Elasticity of Some Ferromagnetic Substances. K. Nakamura. (Science Reports of the Tôhoku Imperial University, 1935, vol. 24, Aug., pp. 303–331). Three methods for measuring Young's modulus in ferromagnetic substances, and its variation ($\triangle E$) by magnetisation, are described. Using these methods the $\triangle E$ effect for iron, nickel, cobalt and nickel-iron alloys was measured, and the following results were obtained: Young's modulus for iron and cobalt was almost independent of the magnetisation; for nickel it was found to be proportional to the magnetisation. The $\triangle E$ effect is very small for nickel-iron alloys with less than 30 per cent. of nickel; it reaches a maximum at 50 per cent. of nickel, but is generally smaller than for pure nickel.

Comparison of Nickel and Nickel-Chromium Steels with

Plain Carbon Steels. A. Leblanc. (Métaux, 1935, vol. 10, May, pp. 131–134). Three diagrams are presented in which the relationships of the elastic limit to the notch-toughness, that of the breaking strength to the notch-toughness and that of the breaking strength to the elastic limit, for plain carbon, and nickel and nickel-chromium steels are plotted. The compositions of the materials to which the plottings refer are not stated.

The "Oxford" Impact Machine. (Engineering, 1935, vol. 140, July 19, pp. 54, 55). In the "Oxford" machine both anvil and striker are suspended by wires so that no shock is transmitted to earth, and four-point loading is employed. The correlation between figures obtained by tests on the Oxford machine and the standard Izod apparatus is very good. The machine described is of the same type as that previously constructed, but it is larger, and minor refinements have been made. (*See* Journ. I and S.I., 1934, No. II., p. 652).

Standardisation of a Small Notch-Impact Specimen. R. Mailänder. (Stahl und Eisen, 1935, vol. 55, July 11, pp. 749–754; July 18, pp. 779–785). In view of the necessity for a standardised test-piece for carrying out notched-bar impact tests the Deutsche Verband für die Materialprüfungen der Technik and the Verein deutscher Eisenhüttenleute appointed committees to consider the choice of a suitable small test-piece. In the present article the author gives a detailed account of the results of the researches of these committees. After discussing the nature and features of the notched-bar impact test, the information which it yields, the need for standardised specimens, and the general points of view to be taken into account in selecting the dimensions of the test-piece, the author records the results of impact tests made at 20° C. on specimens of various widths and depths, with notches of different depths and sharpness, to distinguish between steels which had received correct or incorrect heat treatments. The differentiation between the individual specimens and the scattering of the results, and also the relationships between the notch-toughness values given by them are discussed. On the basis of the results obtained and the considerations enumerated it is suggested that the standard small specimen should measure $10 \times 10 \times 55$ mm. and that the notch should be 3 mm. deep and 2 mm. in diam.; as a supplementary test-piece, one having a notch only 1 mm. in diam., but otherwise identical with the first, is proposed. The differences between the standard small specimen advocated and those generally used in other countries outside Germany are indicated.

Standardisation of the Notched Bar Test Piece. (Metallurgist, 1935, Oct. 25, pp. 75–77). An English summary of the views of A. Steccanella (La Metallurgia Italiana, 1935, vol. 27, Feb.,

pp. 81–108) and of R. Mailänder (see preceding abstract) on the most suitable type of test-piece for use in notched-bar impact determinations is given.

The Notch Sensitiveness of Steel under Alternating Stress. R. Mailänder. (Technische Mitteilungen Krupp, 1935, vol. 3, No. 3, pp. 103–111). The concept notch sensitiveness (η_k) had been introduced in the hope that it would be found to be a constant of the material, that is, that for a given material it would have the same value, whatever the shape of the notch. It is calculated from the formula :

$$\eta_k = \frac{\beta_k - 1}{\alpha_k - 1}$$

where β_k is the ratio of the fatigue strength of an unnotched specimen to the nominal fatigue strength of a notched test-piece, and α_k is the ratio of the maximum stress at the root of the notch in a specimen, set up by a load which nowhere exceeds the elastic limit, and the nominal stress for that same load. The author's experiments show that this notch sensitiveness index (η_k) is not suitable for revealing small differences of quality between various steels, because it is markedly influenced by a very slight scatter of the experimental results, and internal notch action can give the appearance of a low notch sensitiveness. No relation between the notch toughness and the notch sensitiveness could be observed.

Report on Effect of Long-Time Heating With and Without Stress on Impact Resistance of 18 per cent. Chromium, 8 per cent. Nickel Steel (K9 and K19 Steel). (American Society for Testing Materials, 1935, Preprint No. 24, pp. 6–12). Cast and wrought 18 per cent. chromium, 8 per cent. nickel steel (K19) showed moderate loss of impact resistance after heating for 100 hr. at 1,400° F., and only minor changes in impact resistance with further heating up to 3,870 hr. A satisfactory impact resistance remained after 3,870 hr. heating at 1,400° F. The K19 steel first showed, after 1,000 hr. heating at 1,400° F., a lower percentage reduction than that shown by the K9 steel in previous tests but ultimately this approached the same value. Heating at 1,400° F. for 100 hr. under stress produced little loss of impact resistance for the two steels. The data presented suggest that embrittlement, measured by impact resistance, may go on at a slower rate in a stressed than in an unstressed condition.

Impact Resistance of Certain Nickel Steels at Low Atmospheric Temperatures. D. A. Campbell. (Transactions of the American Society for Metals, 1935, vol. 23, Sept., pp. 761–781). The author has studied the influence of nickel in steel in small amounts and the influence of heat treatment on pearlitic nickel steels with respect to impact resistance at low atmospheric temperatures.

Heat treatment, other than normalising or annealing, increases the impact resistance at low temperatures of plain carbon and pearlitic nickel steels. A maximum impact resistance at ordinary atmospheric temperature ranges is obtained when the steels are quenched and fully drawn. The addition of nickel in small quantities tends to improve the impact resistance of normalised low-carbon steels at low temperatures. It is indicated that the amount of nickel required to give the maximum low atmospheric temperature value is a function of the carbon content of the steel. With less than 0·4 per cent. of carbon, maximum impact resistance at low atmospheric temperatures is obtained with about 2 per cent. of nickel. With carbon in excess of 0·4 per cent. maximum impact resistance is obtained with at least 3·5 per cent. of nickel. The addition of about 2 per cent. of nickel to a low-carbon steel retains the room temperature impact resistance down to – 35° F. if the steel is in the normalised condition. In this respect the addition of this amount of nickel has the same effect as quenching and tempering a plain carbon steel to give a completely sorbitic structure.

Interpretation of Torsion Impact Properties of Carbon Tool Steel. G. V. Luerssen and O. V. Greene. (American Society for Metals, Oct., 1935, Preprint No. 24). The characteristic torsion impact curve of quenched and tempered tool steel presented by the authors in previous papers (Proceedings of the American Society for Testing Materials, 1933, vol. 33, p. 315, and Transactions of the Society for Metals, 1934, vol. 22, p. 311) is discussed. The peculiarities of this curve can be explained by the structural and physical changes occurring in the tempering operation. Particular reference is made to the drop in impact resistance between tempering temperatures of approximately 175° and 230° C. It is shown that this drop in impact resistance is accompanied by a decrease in ductility, and by means of reconstructed impact curves calculated from hardness and ductility measurements, this decrease in ductility is proved to account for the drop in impact value. The reduction in ductility, and consequently in impact value, is caused by a structural change involving the decomposition of austenite. It is concluded that the first, or rising, stage of the curve is the result of stress relief in quenched martensite, the second, or falling, stage is caused by the breakdown of retained austenite to the less ductile martensite, and the third, or rising, stage results from the gradual breakdown of martensite into the softer more ductile constituents.

Photo-Elastic Tests Require Care in Preparing Specimens. T. V. Buckwalter, O. J. Horger, and S. Maulbetch. (Steel, 1935, vol. 96, June 3, pp. 31–32). The authors, who are working on a programme of photo-elastic tests for the Timken Roller Bearing Co., describe briefly the apparatus set up for the study of press

fits; they also indicate the care which must be employed in preparing the specimens. The bakelite specimen is cut, oversize, from sheet, and annealed in an electric furnace at 80°-95° C. The specimen is then carefully machined to size, and again annealed.

The Use of the Hele-Shaw Apparatus in the Investigation of the Flow of Metals. A. M. Herbert and F. C. Thompson. (Iron and Steel Institute, Sept., 1935: this Journal p. 117).

Some Residual Stress Effects in Static Tension Specimens. R. E. Jamieson. (Journal of the American Welding Society, 1935, vol. 14, Sept., pp. 16–21). Experiments were carried out in order to measure the effect of residual stress on the behaviour of welded specimens under tension. Strips of mild steel, 5 ft. × $4\frac{3}{4}$ in. × $\frac{5}{8}$ in., were grooved on the edges, stress-relieved at 1,150° F. and then welded. Measurements of the welding strains and residual stresses were obtained, using Martens extensometers. Welded and unwelded specimens were tested in a tensile testing machine, and the stress/strain curves were compared. It was found that the presence of residual welding stresses exerts a considerable effect on the shape of the stress/strain curve. The high local stresses give rise to larger deformations than those exhibited by the unwelded plate, but the overall strength is not seriously affected. Welding raises the yield point and lowers the ductility, and reduces the ultimate strength by about 9 per cent.

An X-Ray Study of Lüders Lines and Strain Figures in Low Carbon Steel. R. F. Mehl and M. Gensamer. (Metals and Alloys, 1935, vol. 6, June, pp. 158–159). X-ray photograms from strained and unstrained samples of automobile sheet steel showed that the unstrained specimens yielded Debye rings made up of discrete spots, whereas specimens which showed Lüders lines or strain figures gave Debye rings of nearly uniform intensity. It was not possible to distinguish pre-strain in this way, however.

The Piobert Effect in Iron and Soft Steel. E. W. Fell. (Iron and Steel Institute, Sept., 1935: this Journal, p. 75).

Stresses in Overstrained Materials. (Report of Committee to Consider Stresses in Overstrained Materials, British Association, Sept. 5, 1935: Engineering, 1935, vol. 140, Sept. 13, pp. 291–292). The Committee consider that the yield point affords a more reliable basis for design than the elastic limit, but, like the latter, it also is open to objections concerning the method of testing. The "lower yield point" is suggested as being the most trustworthy basis of comparison for the different structural steels.

Magnaflux Inspection for Cracks and Seams. (Iron Age, 1935, vol. 136, Sept. 26, pp. 40–41, 143). In the Magnaflux method of testing ferrous materials for cracks, seams and hidden discontinuities, the part to be tested is magnetised and dusted

with Magnaflux powder; the latter reveals the location of any such defects, if present. Some applications of this system to bars, crankshafts, &c., are referred to.

Flash Magnetisation for Examining Castings and Forgings. A. G. Warren. (Engineering, 1935, vol. 140, Oct. 4, pp. 353-354). Defects in steel castings or forgings may be revealed by magnetising the specimen and coating it with finely divided iron dispersed in a suitable medium. Flaws are thus rendered visible. The method is somewhat difficult to apply to rotor forgings, turbine discs, pressure cylinders, &c., owing to their size and form. However, these articles may be circularly magnetised by winding a coil of wire through the bore and passing a current through the winding. Since the magnetic circuit is a closed one over 90 per cent. of the magnetisation is retained on switching off the current; hence this need only be maintained for a short time, depending on the rate of growth. Eddy currents induced in the specimen reduce the back e.m.f. and permit a very rapid rise to the maximum current value. With flash magnetisation the current is maintained for a sufficient length of time to blow a fine fuse. The loop of conductors used for magnetising the specimen usually consists of a "rope" of conductors connected to switches at each end, the whole being the equivalent of a coil of eighteen turns and a resistance of 4 ohm. The instantaneous current attained is of the order of 100 amp. at "blow." This suffices for the effective magnetisation of a tube 2 ft. in diam. and 24 ft. long; larger articles may be dealt with by using extension loops.

Age Hardening of Metals. H. M. Boylston. (Industrial Heating, 1935, vol. 2, Jan., pp. 9–13). The author reviews the explanations which have been advanced concerning age-hardening and the effects of carbon, nitrogen, cobalt and molybdenum upon the ageing phenomenon.

Age Hardened Steel Avoided rather than Utilised. P. D. Merica. (Metal Progress, 1935, vol. 27, June, pp. 29–32). It is pointed out that the age-hardening effect in mild steels has been known for some time. It can be brought about by carbon and certain other elements, notably copper. Thus a copper steel is susceptible to three types of hardening: Martensitic hardening, carbon · precipitation-hardening, and copper precipitation-hardening. Ageing in steels appears to be related to other phenomena, such as temper-brittleness, blue-brittleness, and work-hardening.

The Ageing of Steel. E. S. Davenport and E. C. Bain. (American Society for Metals, Oct., 1935, Preprint No. 9). The authors deal first with the moderate increase in hardness which develops gradually in annealed low-carbon steel at temperatures as low as room temperature following a rapid cooling from temperatures near 1,300° C. This is explained on the basis of

the precipitation of iron carbide from ferrite supersaturated with respect to carbon, a reaction which conforms in type with those in familiar age-hardening alloys. More extended studies are reported on the ageing phenomena observed at temperatures from 0° C. upwards in cold-worked low-carbon steel. Observations from many sources are considered, which, as summarised, indicate that strain-ageing, blue-brittleness, the Fry etching figures and other effects are all due to the precipitation, particularly upon slipbands, of a compound, probably an oxide derived from ferrite supersaturated with respect to oxygen.

The Influence of Deoxidation on the Ageing of Mild Steels. B. N. Daniloff, R. F. Mehl and C. H. Herty, jun. (American Society for Metals, Oct., 1935, Preprint No. 8). A series of steels, varying in carbon content between 0·02 and 0·25 per cent. and representing a wide variation in the degree of deoxidation, was tested for ageing after quenching from subcritical temperatures and after cold working. According to the deoxidation treatment the steels were divided into four major classes: (1) Rimmed steels, (2) semi-killed steels, (3) silicon-killed steels, and (4) aluminium-killed steels. It was found that deoxidation had a very pronounced effect on both types of ageing. The tendency of steels to age was found to decrease with deoxidation in the order given above. Steels deoxidised with silico-manganese in the furnace and ferro-silicon and aluminium in the ladle were found to be practically non-ageing. It is believed that deoxidation decreases the susceptibility of steels to ageing directly through the decrease in the dissolved oxygen content, and indirectly by virtue of its action in refining the grain-size of the steels.

Magnetic Ageing of Iron due to Oxygen. T. D. Yensen and N. A. Ziegler. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 624). The authors explain that the phenomenon of magnetic age-hardening in iron has hitherto been attributed to the presence of nitrogen and carbon; oxygen has not been considered to exert an important influence. The authors prepared ring samples of electrolytic iron melted under hydrogen in a H.F. furnace, and tested them magnetically. Oxygen (above that already present) was introduced by heating the samples in air at 700°–900° C. and air-quenching; the oxygen contents were 0·019, 0·029 and 0·035 per cent. for the three specimens used. The magnetic properties were again obtained. Each sample was then aged by heating *in vacuo* for 1 hr. at temperatures of 200°, 400°, 600° and 800° C., followed by slow cooling and magnetic testing. From the results thus obtained the authors conclude that precipitation hardening occurs owing to the presence of oxygen, and that there is probably a critical particle size giving rise to the greatest magnetic effect.

Influence of Nitrogen and Oxygen on the Age-Embrittlement

of Steel. (Metallurgist, 1935, Aug., pp. 52–53). This is an English summary of a paper by W. Eilender, H. Cornelius, and H. Knüppel, published in Archiv für das Eisenhüttenwesen, 1935, vol. 8, May, p. 507. (*See* Journ. I. and S.I., 1935, No. I. p. 469).

Correlation of Failures from Embrittlement of 4 to 6 per cent. Chromium Steel with the Notched Bar Impact Test. H. M. Wilten. (American Society for Metals, Oct., 1935, Preprint No. 17). Chromium steel tubes (chromium 4 to 6 per cent.) when used in refinery cracking stills in the form of coils are subject to embrittlement. The author has studied the cause of this embrittlement and has correlated the results obtained from service data and laboratory tests. Chromium steel tubes containing molybdenum did not exhibit embrittlement in service.

The Embrittlement of Boiler Steel.—I. E. P. Partridge and W. C. Schroeder. (Metals and Alloys, 1935, vol. 6, June, pp. 145–149). In this, the first part of a correlated summary on the subject of caustic embrittlement, the chemical factors in the attack of boiler waters upon steel are considered. The ionic reactions involved are discussed, together with the effect of dissolved substances, notably sodium hydroxide. From data which they adduce, the authors contend that the oxide coating normally formed by chemical action acts as a barrier to further penetration, and that in certain cases the presence of salts in solution may tend to dissolve the oxide coating and so expose the metal to chemical attack. Discontinuities in the oxide film are likely to accelerate localised attack.

The Embrittlement of Boiler Steel.—II. E. P. Partridge and W. C. Schroeder. (Metals and Alloys, 1935, vol. 6, July, pp. 187–191). Attention is directed to the effect of chemical attack on the mechanical properties of low-carbon steel. The findings of numerous investigators on hydrogen embrittlement after pickling are quoted, and the subject of caustic embrittlement is subsequently treated. Among the conclusions drawn by the authors are the following: At boiler temperatures the solubility of hydrogen in iron is very low, but it may be greatly increased by chemical reaction at the surface of the metal; the hydrogen diffuses out from the metal on the cessation of chemical reaction; steel which has absorbed hydrogen behaves in a brittle manner, the hydrogen content necessary to cause brittleness being quite low; recovery from embrittlement is usually incomplete; the likelihood of embrittlement by an aqueous solution depends largely on the degree to which the solution removes or prevents the formation of the protective coating of oxide on the metal; selective corrosion of steel subjected to repeated or reversed stresses may lead to fatigue failure.

The Embrittlement of Boiler Steel.—III. E. P. Partridge and W. C. Schroeder. (Metals and Alloys, 1935, vol. 6, Sept., pp.

253–258). After reviewing the properties of low-carbon steel at boiler temperatures and the part played by ageing in giving rise to embrittlement, the authors discuss the cracking of low-carbon steel under combined chemical attack and mechanical stress. Reports of failures in vessels containing hot caustic solutions are examined, and these failures are compared with embrittlement failures in boiler steel.

Boiler Steel Embrittlement. E. P. Partridge and W. C. Schroeder. (Mechanical Engineering, 1935, vol. 57, May, pp. 294–296).

On the Cold Shortness in Metals and Alloys. Y. Fujii. (Kinzoku No Kenkyu, 1935, vol. 12, Apr. 20, pp. 189–209). The author has measured the shock resistance of thirty-seven pure metals and alloys of practical importance, within the temperature range – 190° to 500°. From the results the following conclusion is drawn : Cold-shortness depends mainly on the type of crystal space lattice ; metals with a face-centred cubic lattice are free from cold-shortness, while metals with other types of lattice reveal an increase in brittleness at low temperatures.

A New Method and Machine for Dynamic Hardness Testing. W. M. Patterson. (American Society for Testing Materials, 1935, Preprint No. 38). The author describes the Ballentine dynamic hardness tester, and explains the development of a scale fulfilling the desired conditions. Test data are given on alloys of lead, copper, and aluminium, cast iron and soft and medium-hard steels. The reproducibility of results to be expected with the method in laboratory and production work is estimated, and a brief discussion of the advantages and particular field of use of the machine is given.

Hardness Testing of Nitrided Steels. J. H. Hruska. (Iron Age, 1935, vol. 136, Sept. 26, pp. 22–23, 26–28, 30, 58). In view of the hardness of nitrided surfaces and the thinness of the nitrided case the ordinary methods of determining case-hardness are difficult to apply. The comparative merits of the various types of hardness testing machines when used on nitrided material are discussed.

Wear and How to Determine it. W. A. Harrington. (Steel Treatment Research Society, Australia : Heat Treating and Forging, 1935, vol. 21, May, pp. 233–237). It is pointed out that wear resistance is not synonymous with hardness, and a special machine for testing resistance to wear is described. The machine consists of a rotating disc carrying an annular track which may be supplied with an abrasive or lubricant, as desired. The test specimens are discs of known radius (9 mm.) and width (6 mm.), and are mounted in pairs in a pivoted holder. During the test the discs rub against the rotating track, thus producing a flat on the specimens. The length of this flat is determined by a micro-

scope, and the volume of material worn off is calculated. The load on the specimens may be adjusted to give the required bearing pressure. Results of tests on cast iron for piston rings (using a cylinder iron on the rotating disc and paraffin as the lubricant) show that hardness is not related to wear resistance, and that a pearlitic structure is the most resistant. In these tests the wear-resistance was determined by comparing the rate of wear with that of a standard specimen tested at the same time.

Reducing Abrasion by Compound Contact Pieces. S. Saito and N. Yamamoto. (Metal Progress, 1935, vol. 27, June, pp. 52–55). Apparatus was set up by the authors which consisted of a revolving railway tyre upon which two specimens were allowed to rub at predetermined pressures. It was discovered that a considerable reduction in wear could be effected if the specimens were of dissimilar materials. Thus, if the test-pieces were of steel and cast iron (bearing on a steel tyre) the wear on the steel specimen was reduced to a negligible amount; it was as if the cast iron exerted a lubricating action. The reduction in wear depends largely on the rubbing pressures; by adopting a sufficiently low bearing pressure for the cast-iron specimen the wear on the latter is diminished, whilst the lubricating action is maintained. Similar tests on steel allied with other materials, such as teak, resin, and asbestos, revealed results of the same type. These findings have been applied to railway practice with considerable success. Experiments were also carried out to investigate the wear of collector shoes on trolley wires. By using an abrasive pair, consisting of a copper shoe and a shoe of hard rubber, the wear on both the trolley wire and the copper collector was reduced almost to zero. The authors suggest that the above phenomena may be explained in the following way: The abrasion of a ductile material, such as steel upon steel, gives rise to considerable roughening of the surface. If cast iron is used the brittle particles of cast iron become embedded in the rubbing surfaces, and these therefore become smooth. A soft material such as tin or lead may be considered to smear the surface and produce a similar lubricating effect.

The So-Called "Bleeding" [*Bluten*] of Steel. A. Leon. (Montanistische Rundschau, 1935, vol. 27, Stahlbau-Technik, July, pp. 1–3). On taking down steel parts which have been subjected to alternating stresses while joined together, a brownish-red coating (called *Edelrost* or *Passungsrost* in German) may be found on the surfaces that had been pressed together. The phenomenon is called "bleeding" (*Bluten*) or "sweating" (*Schwitzen*), and it occurs not only on structures which are apparently immovably fixed together, but also on parts which slide or roll over one another (plain and roller bearings). Chemical and X-ray examinations have shown the "rust" to consist of

ferric and ferroso-ferric oxides mixed with metallic particles; as the iron has combined directly with oxygen (without the presence of moisture as required for ordinary rusting), the term "frictional oxidation" has been proposed. The present author discusses the discovery of frictional oxidation, its independence of the temperature, the cause of it, and its connection with wear and with fatigue fracture, on the basis of the publications of several investigators.

The Cutting of 18% Chromium, 8% Nickel Steel with Silver-Steel Tools. W. Ulfert. (Stahl und Eisen, 1935, vol. 55, Oct. 3, p. 1070). Special-steel tools are usually considered necessary for the machining of rustless chromium-nickel steel. The author's tests with silver-steel tools, containing 1·2 per cent. of carbon and 1 per cent. of tungsten, show that this material can well be used for the tools, provided that a cutting angle not usually employed is adopted; cooling with cutting oil is essential. Diagrams of suitable tool-shapes are given.

Factors Influencing the Nature of the Cutting Speed—Tool Life Curve. O. W. Boston, W. W. Gilbert and C. E. Kraus. (American Society for Metals, Oct., 1935, Preprint No. 26). The object of this paper is to show that the exponent, as well as the constant, in the formula $VT^n = C$ may vary as conditions are changed. It is shown that the relation between the cutting speed and tool life conforms to the above mathematical equation, in which: V is the cutting speed in ft. per min.; T is the tool life under cut up to the time of tool breakdown, measured in minutes; n is the exponent of T and equals the tangent of the angle of slope of the cutting-speed/tool-life curve when plotted on log-log paper; and C is a constant depending upon the various conditions. In order to show the influence of each of several factors, such as the tool material, the tool shape, the material cut, and the cutting fluid, the results of several sets of tests are presented and discussed. It is shown that a variation in any of these factors may change the value of n as well as the constant C in the above equation.

Grinding Cracks. A. M. Steever. (Metal Progress, 1935, vol. 27, May, pp. 52–53, 68). Articles such as dies, which during manufacture are subjected to grinding operations, are sometimes found to crack during or soon after grinding. Such cracks have a characteristic mosaic appearance; they are the result of harsh grinding. The cracks may be revealed by boiling for 15–30 min. in a solution of hydrochloric acid and water, or by pickling in nitric acid. Grinding cracks may be avoided if care is taken in the final grinding operation, using light cuts and an ample supply of cooling liquid.

Cold-Work in the Boiler Shop. N. Dewulf. (Revue de Métallurgie, Mémoires, 1935, vol. 32, May, pp. 215–224). The

author reports briefly tests made to determine the influence of cold-work, applied by tension, by bending and by compression, on the properties of normalised boiler-plate material and the effect of heat treatment on the changes produced.

Electro-Magnetic Tests for Work-Hardening Steel Wires. T. F. Wall. (Engineer, 1935, vol. 160, Sept. 13, pp. 260–263). The purpose of the investigations was to provide means of observing the changes in the elasticity and viscosity of steel wires of colliery winding and haulage ropes after they have been in service. Two methods were employed: (1) A ballistic magneto-striction method which consists essentially in magnetising the wire by a known magnetising force and then applying a load to the wire; the increase of induction in the wire due to the application of load is measured by the throw of a ballistic galvanometer; (2) the observation of a travelling surge of stress in the wire; the wire is surrounded by solenoids connected in series, and, by means of a valve amplifier, to an oscillograph; a blow is struck at one end of the wire, and the surge of stress so induced travels along the wire and gives rise to an e.m.f. in the corresponding search coil, this e.m.f. being recorded on the oscillograph. Tests carried out by the magnetostriction method showed not only considerable work-hardening in used ropes, but also evidence of the same effect with new ropes. Surge stress results confirmed this, and also revealed that the value of Young's modulus for old wire is reduced to a very low figure, suggesting considerable deterioration. Data relating to surge stress tests on Armco wire and on pure nickel wire are also given.

An Apparatus for Measuring the Thermal Conductivity of Metal Sheets. A. Eucken and H. Warrentrup. (Zeitschrift für technische Physik, 1935, vol. 16, pp. 99–104). The apparatus described functions on the principle that heat is imparted to the middle of the sheet to be tested while the edges are kept at a constant temperature, the constant temperature difference so set up being then measured.

The Shape of the Stress-Elongation Curves of Steel in the Blue-Heat Temperature Range. W. Enders and W. Lueg. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1935, vol. 17, No. 6, pp. 77–88). Tensile tests were carried out in the blue-heat range on a number of low-carbon steels; the stress was measured by an electrical device incorporating the principle of an electric condenser. By this means the influence of various heat treatments on the number and magnitude of the sudden deviations in the load-extension curves, the yield points, the tensile strengths, and their dependence on the temperature were determined. The sudden deviations in the extension curves, entailing corresponding decreases of load, occurred only in a narrow range of temperature; for the steels tested in the as-received condition,

this lay between 100° and 280° C. for the rates of extension used. In all the curves relating the temperature to the number of jumps there was a clearcut maximum, the corresponding temperature of which seemed to be dependent on the nitrogen and oxygen contents of the steel, and so on its mode of manufacture. In steels containing much gas, the number of jumps and the corresponding temperature could not be altered by a long low-temperature anneal nor by extremely slow cooling from 680° C., but in steels with lower oxygen and nitrogen contents such treatment lowered the maximum number of jumps and shifted it to higher temperatures —in one case, a long low-temperature anneal almost completely suppressed the jumps. Special heat treatment of a hot-rolled soft iron increased the number of jumps, but the magnitude of the individual deviations became almost imperceptible. Whether these jumps are connected with precipitation phenomena cannot be stated; they occurred in all the steels in the as-received condition, the micrographs of which revealed no precipitation, and also after heat treatment, both in steels showing marked precipitation and in others with none. In duralumin (on which tests were also made) the occurrence of similar deviations appeared to be related to precipitation phenomena. With the appearance of jumps in the blue-heat range, an increase in the tensile strength was observed; in all the steels the maximum in the number of jumps occurred at a somewhat lower temperature than did that of the tensile strength. Slow cooling from 680° and prolonged annealing at 200° C. both lowered the yield point and the tensile strength, but the notch toughness was lowered only in the case of steels containing much gas; it was noteworthy that the marked decrease of the notch toughness occurred without previous cold-deformation.

The Influence of High Temperatures on the Strength Properties of Soft and Medium-Hard Steels under Static and Dynamic Stressing. K. Uhlemann. (Metallwirtschaft, 1935, vol. 14, Sept. 27, pp. 773–778; Oct. 4, pp. 795–798). The changes in the static and dynamic strength properties due to high temperatures were investigated in four carbon steels of 40 to 80 kg. per sq. mm. strength, and one low-alloy steel, at temperatures up to 500° C. The tests included hot tensile, bending fatigue and notched-bar impact tests; in the latter, particular attention was paid to the influence of the depth of the notch. The results obtained are set out in tables and diagrams; the influence of scaling on the bend fatigue strength, the relation between the static and dynamic properties, the types of fracture, and the practical conclusions to be drawn from the tests are discussed.

Investigation of the Behaviour of Metals under Deformation at High Temperatures. Part I.—Structural Changes in Mild Steel and Commercial Irons during Creep. C. H. M. Jenkins

and G. A. Mellor. (Iron and Steel Institute, Sept., 1935: this Journal, p. 179).

The Behaviour of Mild Steel under Prolonged Stress at 300° C. C. H. M. Jenkins. (Iron and Steel Institute, Sept., 1935: this Journal, p. 281).

Mechanical Study of Metals at Raised Temperatures. P. Chevenard. (Métaux, 1935, vol. 10, Mar., pp. 76–86). The author describes and illustrates the apparatus installed in the laboratories of the Imphy Steelworks for carrying out tests on metals (bars and wires) at raised temperature. These include an "isothermal viscosimeter," in which the "flow" of a loaded wire at a constant raised temperature is measured; a "viscosimeter for rising temperatures," which is a dilatometer for recording the flow of a loaded wire as a function of the temperature; "relaxometers," in which the specimen (wire or rod) is subjected to a load by means of a spring, and the relaxation, or decrease of load that sets in under the influence of the temperature, is measured; a "torsion oscillometer"; and an industrial tensile testing machine for raised temperatures. The types of curves obtained by the machines are illustrated.

Creep Tests by the Method of W. Rohn. R. Scherer and H. Geipel. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Aug., pp. 99–102). In Rohn's creep testing machine the expansion of the specimen is made to control the furnace temperature, so that the creep under the test load and the thermal expansion are in equilibrium, that is to say, the specimen length varies about a constant value during the test. The applied load is then the creep limit for the temperature at which equilibrium is finally established. The present authors have carried out an investigation of this test method; they have examined the temperature distribution in the testing machine, the influence of the specimen length on the sensitivity of temperature control, the dependence of the equilibrium temperature corresponding to the creep limit, as determined, on the temperature at the start of the test, and the influence of the coefficient of expansion and of the elastic modulus of the test material on the creep. They discuss the determination of the rate of extension from the time/temperature curve and the influence of inaccurate temperature measurement on this calculation. Finally, they make a comparison of creep values obtained by the Rohn method with the creep strength determined in a test at constant temperature and under constant load on the basis of a rate of extension of 5×10^{-4} per cent. per hr. in the period from the twenty-fifth to the thirty-fifth hour.

Influence of Time on Creep of Steels. A. E. White, C. L. Clark, and R. L. Wilson. (American Society for Testing Materials, 1935, Preprint No. 32). The results are presented of extended-time creep tests carried out at 1,000° F. on a plain carbon steel of

the S.A.E. 1015 type and a chromium-silicon-molybdenum steel. Certain of the stresses for the alloy steel were so selected that fracture would result at various periods of the test, thus producing the complete time-elongation curves. These stresses were considerably greater than any which would normally be used in commercial practice. The fractured specimens were subjected to metallographic, X-ray and hardness examinations to determine the extent to which the structural and deformation characteristics were dependent on the time required for rupture. The results presented indicate that the creep ratio, and thus the reported creep stresses, are influenced by the length of time to which the creep tests are extended. While not affected to the same degree, with both steels lower creep strength values were obtained from the 500-hr. test than from the 1,000-hr. test, and the values obtained were more comparable with those obtained from tests of several thousand hours' duration. The structural and deformation characteristics were found to be influenced by the time required for rupture. The observed differences may be explained either on the assumption that the mechanism of deformation is different in the short-time tension test from that in the creep tests, or that, with the creep specimens, the deformation characteristics were changed because of the length of time at temperature.

Stability of Steels under Stress up to 1,000° F. E. L. Robinson. (Metal Progress, 1935, vol. 28, Sept., pp. 34–39, 78). The author commences by pointing out that in turbine work the permissible creep is very low—of the order of a few hundredths of 1 per cent. over a period of years. Attention is then directed to the various forms of time/creep diagrams and the facts which they reveal; these diagrams are analysed with particular reference to the considerations which are paramount in turbine practice.

Some Alloys having Great Thermal and Magnetic Stability. P. D. Merica. (Metal Progress, 1935, vol. 28, July, pp. 41–45). Certain alloys of the age-hardening type possess a high mechanical and thermal stability at elevated temperatures, after age-hardening; "Konal," containing nickel about 70, cobalt 17, iron 6, and titanium 2 per cent., and a recently-developed tungsten-cobalt-iron alloy containing tungsten 30, cobalt 25, and iron 30 per cent., are of this type. The author remarks on these alloys, and on the alloys which exhibit an analogous magnetic hardness (high coercive force). The tungsten-cobalt magnet steels and aluminium-nickel-iron alloys are cited in this connection. Finally, the author comments on certain discrepancies and inconsistencies in the age-hardening phenomena commonly observed.

Influence of Carbon Content on High Temperature Properties of Steels. A. E. White, C. L. Clark, and R. L. Wilson. (American Society for Metals, Oct., 1935, Preprint No. 18). The authors have investigated the influence of the carbon content on the high-

temperature characteristics of two types of steel, one a plain carbon steel and the other a low-alloy steel containing silicon, chromium and molybdenum. Two carbon contents were selected for each type, the lower being approximately 0·10 per cent. and the higher 0·45 per cent. Both the short-time tensile properties and the creep characteristics were studied. The tensile tests were carried out at temperatures ranging from 26° to 760° C., and the creep tests at 425°, 540° and 650° C. The impact resistance was also studied. The relative high-temperature stability of the steels was also investigated by submitting specimens after creep testing to metallographic examination and to tensile and impact tests at room temperature. The results indicate that the behaviour of steels at elevated temperatures is sometimes independent of the carbon content and may be affected by initial heat treatments or by the addition of alloying elements that would change the character of the carbides in the structure.

High Temperature Properties of Nickel-Cobalt-Iron Base Age-Hardening Alloys.—Part I. C. R. Austin. (American Society for Metals, Oct., 1935, Preprint No. 20). The author presents the results of a study of the properties of nickel-cobalt-iron alloys. The mechanical properties, age-hardening characteristics, work-hardening and temperature softening properties of the alloys are considered. The hardest alloy of the series investigated was found to maintain its hardness (590–635 Vickers) at 1,110° F.

Physical Properties of Metals as Affected by Conditions of Ammonia Synthesis. H. L. Maxwell. (American Society for Metals, Oct., 1935, Preprint No. 16). The problems to be considered in the selection of materials for such applications as high-pressure hydrogenation and ammonia synthesis are tensile strength, ductility and shock resistance at high temperatures, and resistance to hydrogen attack and creep under long continued high stresses. The work reported was undertaken in order to gain further information on metals having the desired properties for these operations. The results show that deoxidised copper and copper alloys are more resistant to high-temperature, high-pressure hydrogen-nitrogen mixtures than similar alloys containing oxide. Plain carbon steel and also low-chromium nickel alloys are severely embrittled at temperatures up to 500° C. in contact with 3 : 1 hydrogen-nitrogen mixtures at 1,000 atm. pressure. The chromium-tungsten, chromium-molybdenum, and chromium-vanadium steels show increased resistance to gas attack under these same conditions for intervals up to 1,535 hr.

The Rotating-Wire Arc Fatigue Machine for Testing Small-Diameter Wire. J. N. Kenyon. (American Society for Testing Materials, 1935, Preprint No. 35). A stress-reversal fatigue machine developed on the principle of a rotating wire bent to arc-

curvature is being used to test small-diameter wire. Resonant vibrations or standing waves are damped in an oil bath. The specimen automatically assumes the form of a circular arc by the elimination of flexural shear. Test results so far obtained give satisfactory values.

Fatigue Testing Machine for Wire. (Wire and Wire Products, 1935, vol. 10, July, pp. 272–274, 284, 285). The Haigh-Robertson fatigue testing machine for wire is described. (*See* Journ. I. and S.I. 1934, No. II. p. 676).

Some New Fatigue Testing Machines. J. S. G. Primrose. (Transactions of the Manchester Association of Engineers, 1934–1935, pp. 147–152). Short notes on fatigue testing machines of recent origin are given.

An Electromagnetic Fatigue Tester. (Engineering, 1935, vol. 140, Oct. 11, p. 406).

New Fatigue Tester. (Iron and Coal Trades Review, 1935, vol. 131, Aug. 23, p. 275). In this apparatus—developed by Salford Electrical Instruments, Ltd.—the sample bar is supported at its two nodes and vibrated electromagnetically so that it resonates. By measuring the deflection at an antinode the applied stress can be calculated. When fracture occurs a trip is operated and the machine stops. The tester is designed for steel bars 18 in. long, though other sizes may be dealt with.

Notes on Fatigue Tests on Rotating-Beam Testing Machines. (American Society for Testing Materials, 1935, Preprint No. 25, pp. 3–10). These notes cover tests made on rotating-beam testing machines of either the centrally loaded or the cantilever type.

Effect of Occasional Overload on the Strength of Metals. H. F. Moore. (Metals and Alloys, 1935, vol. 6, June, p. 144). Brief comments on the strength of metals as related to the endurance limit under fatigue and the degree of overstressing are presented.

Damping Capacity. A Factor in Fatigue. G. R. Brophy. (American Society for Metals, Oct., 1935, Preprint No. 12). The damping characteristics and endurance properties of several steels in several conditions of heat treatment were determined and an attempt was made to correlate these properties. The damping capacity appears to be related to the damage resulting to the steel from overstress and notches. High damping steels are resistant to sharp notches, but not to overstress, while low damping steels are resistant to overstress, but sensitive to notches. A limiting damping capacity, it is believed, for different notches is recognised. Below this limit, extreme notch sensitivity occurs. Several factors influencing these properties were studied, including grain-size and structure.

The Damping Capacity of Crankshaft Steels in the Cold and

Hot Condition as Supplied and in Continuous Service. A. Appenrodt. (Mitteilungen aus dem Wöhler-Institute, Braunschweig, 1935, No. 24: Stahl und Eisen, 1935, vol. 55, Sept. 5, pp. 964–965). The results of fatigue tests and determinations of the damping capacity of sixteen steels of various compositions usually employed for the fabrication of crankshafts, are recorded; the tests were made at room temperature, at 75° and at 120°, representing the highest temperature at which crankshafts might be expected to work.

Endurance of Case-Hardened Gears. O. W. McMullan. (American Society for Metals, Oct., 1935, Preprint No. 10.) The author describes the results of dynamometer breakdown tests of rear axle drive gears and pinions. Tests on oil-hardened and several case-hardened steels are given. Comparisons of results are made with physical properties obtained on laboratory test specimens. The conditions of service make it necessary for the pinion to resist softening due to high operating temperatures at the tooth contacts under conditions of partial lubrication, whereas the gear is better cooled and lubricated and requires a strong steel with good wear resistance at lower temperatures. Wear was found to be excessive on medium-carbon, oil-hardened steel. S.A.E. 4615 steels directly quenched gave good results in pinions, and double-treated Krupp and S.A.E. 4820 steels in gears gave the best results among those steels tested.

Calculation of the Notch Fatigue Strength from the Tensile Strength and Reduction of Area. G. Erber. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Aug., pp. 95–96). On the basis of the assumption that both the tensile strength and the deformability must be of influence on the notch fatigue strength the author has developed a formula for the approximate calculation of the notch fatigue strength, in which the latter is equal to the product of the tensile strength and reduction of area (representing the deformability) multiplied by a constant (which varies with the shape of the notch). On checking the formula with data given in the literature it was found that where the deformability factor did not exceed a certain value the notch fatigue limit under alternating bending stresses could be calculated sufficiently accurately. Whether the formula can be used for calculating the fatigue strength of welds remains to be determined; in this case it is suggested that the bending angle obtained in the folding test should be used instead of the reduction of area.

The Load-Deflection Fatigue Test. A Critical Investigation of its Application to Steels. J. W. Cuthbertson. (Iron and Steel Institute, Carnegie Scholarship Memoirs, 1935, vol. 24, pp. 1–50). Numerous tests on five different ferrous materials have been carried out with the object of checking the validity of fatigue determinations based on observations of the alteration in deflec-

tion of a Wöhler specimen subjected to a steadily increasing load. Comparison with the results of endurance tests shows that, for the materials examined, the discrepancy between the true endurance limit and the rapidly found value is of negligible magnitude, provided that the latter test is conducted according to a fixed routine. Important considerations are a preliminary running under load for several hundred-thousand reversals, and a continuous, as distinct from intermittent, load variation. The extent of deviation from linearity which the load-deflection curve shows at the proportionality yield is greater in the case of a tapered specimen than with one of tubular cross-section, owing to the maximum stress in the latter case being concentrated in a relatively small area of the bar. Tests have been carried out at temperatures up to 300° C., and there is no evidence to indicate that the accuracy of the rapid test is adversely affected in consequence of heating of the specimen. By applying a fixed load to the end of the rotating bar and noting the change in deflection with time, the frequency of stress variation being constant, typical curves are obtained, the form of which depends primarily upon whether the stress is above or below the fatigue stress, but not essentially on the material or its temperature. Assuming that the fatigue limit is not exceeded, a point in common with all these curves is the tendency for the deflection ultimately to assume a constant value, although in some cases an initial increase in deflection is followed by a gradual decrease, the progress of which is not entirely arrested even after 10^7 stress reversals. From the point of view of the load-deflection test, the condition of the material is ideal after the specimen has been run under a stress in the vicinity of the fatigue stress for a period of time of sufficient duration for the deflection to reach either a constant or a maximum value. The form of the constant-stress curve suggests a possible alternative method of rapid fatigue determination.

High-Speed Fatigue Tests of Several Ferrous and Non-Ferrous Metals at Low Temperatures. W. D. Boone and H. B. Wishart. (American Society for Testing Materials, 1935, Preprint No. 36). The authors present the results of a number of low-temperature high-speed fatigue tests. Two cantilever fatigue machines, developed at the University of Illinois, using 12,000 r.p.m. series motors, were used in the tests. A description is given of the machines and their operation at temperatures that varied from + 80° F. to −55° F. Unnotched specimens were used to give maximum endurance limits for the materials, while notched specimens were used to show the effect of stress concentration. The endurance limits of duralumin, brass, cast iron, meehanite cast iron, cold-drawn steel, and rail steel were determined at temperatures of + 80° F., + 10° F., −20° F. and −40° F. In

general, as the temperature was decreased the endurance limits of the metals increased. The stress concentration factors showed no consistent change.

The Yield of Steel Wire Under Stresses of Very Small Duration. W. Mason. (Proceedings of the Institution of Mechanical Engineers, 1934, vol. 128, pp. 409–438). A steel wire 53 ft. long was suspended vertically from a fixed point. Near its lower end a stop was soldered to the wire; a tup, sliding on the wire, was allowed to fall on to the stop from appropriate heights. Waves of tensile stress were thus sent up the wire and reflected at the fixed top. Extensions were measured on certain gauge lengths at the top of the wire. The form of the wave front was inferred from these measurements and from the corresponding distances travelled by the stop; the stresses were calculated from the extensions. The impacts of the tup on the stop were repeated after heat treatment, at a normalising temperature, of the middle part of a 4-in. gauge length. The static yield being 16·3 tons per sq.in., a yield was found with a wave rising from 3·4 to 28·7 tons per sq. in., in $1{\cdot}25 \times 10^{-4}$ sec. and falling again to 3·4 tons per sq. in. after a further $0{\cdot}45 \times 10^{-4}$ sec. Yields above the static yield, but lower than the yield just mentioned, were found with waves of less intensity but of longer duration. A limiting "higher" yield stress for indefinitely small duration of the stress is inferred. This limiting stress for the wire as normalised was about 30 tons per sq. in., the maximum stress in static tension being 24·8 tons per sq. in. The results are discussed with reference to the "higher" and "lower" yield stresses.

Influence of Fillets on Stepped Shafts on the Bending Fatigue Strength. E. Lehr and R. Mailänder. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Aug. 17, pp. 1005–1011: Archiv für das Eisenhüttenwesen, 1935, vol. 9, July, pp. 31–35). The present investigation forms part of a larger research, undertaken by the Verein deutscher Eisenhüttenleute and the Verein deutscher Ingenieure, to provide the necessary basic data for more precise calculations in the design of machinery. The authors present the results of alternating bending (rotational) fatigue tests on three steels—one being St50.11, and the others a chromium-nickel-tungsten and a nitriding steel, respectively—the mechanical properties of which differed widely. The results show how great is the influence of the radius of the fillet in stepped shafts; they also demonstrate the effect due to the size of the test-piece in any given range. Other tests now in progress will confirm and extend these findings, but much research remains yet to be done.

The Relation of Fatigue to Modern Engine Design. R. A. MacGregor, W. S. Burn, and F. Bacon. (Transactions of the

North East Coast Institution of Engineers and Shipbuilders, 1934–35, vol. 51, pp. 161–228). This paper is in three parts. In Part I. a number of cases of failure of marine engine parts—notably tailshafts, crankshafts, and so forth—are cited, and it is shown that these failures may be considered to have been initiated by corrosion-fatigue, impact at a stress raiser, heavy impact, or stress concentration arising from a variety of causes. Finally, the published experimental work on the subject of discontinuities is discussed, and it is shown that further work on this subject is necessary. Part II. deals with the problem of design, especially the avoidance of stress-raisers and of corrosion-fatigue. In the last section the evolution of certain modern ideas on elasticity and fatigue is noted. It is pointed out that notch-toughness and high internal damping tend to be related. Reference is made to the size effect and crack spread. A bibliography of 221 references is appended.

The Effect of Heavy Oils and Greases on the Fatigue Strength of Steel Wire. R. Goodacre. (Wire and Wire Products, 1935, vol. 10, Oct., pp. 473–475). Fatigue experiments were performed on wire subjected to the influence of heavy oils and greases, the machine used being a Haigh-Robertson fatigue-testing machine. The fatigue strength of the steel wire was increased by being coated with heavy oil or grease. The fatigue properties increased with increasing viscosity of the oil up to a certain point, after which there was a marked fall. The author is of the opinion that this cannot be solely attributed to the exclusion of oxygen from the surface of the wire, but that the oil (or grease) may have the effect of filling up the surface blemishes and so improving the fatigue strength of the wire. To discover whether this was so, tests were made on polished material. Polishing was found to enhance the fatigue strength, and this was further slightly increased by immersion in oil. Furthermore, the endurance of overstressed material was found to be raised by the presence of oil.

Some Aspects of the Fatigue Properties of Patented Steel Wires. II.—Note on the Effect of Low-Temperature Heat Treatment. E. T. Gill and R. Goodacre. (Iron and Steel Institute, Sept., 1935 : this Journal, p. 143).

Fatigue Tests on a Safety Type Hook. H. J. Gough and D. G. Sopwith. (Engineering, 1935, vol. 140, July 26, pp. 101–102). Results of tests on a special safety hook for cranes are recorded. The essential feature of the hook is a spring tongue which closes the open side of the hook and prevents the sling from becoming detached, or the point of the hook from fouling. In one design of hook sharp angles were introduced where the tongue was housed in the open position. When tested under repeated loading failure occurred at this weak point. The fatigue limit was $3\frac{1}{4}$ tons.

In a modified design the sharp angles were radiused, and the fatigue limit improved to $3\frac{3}{4}$ tons. The tests were carried out in a Haigh electro-magnetic fatigue testing machine of 6 tons capacity, using cycles of repeated unidirectional loading.

The Examination of Bridges from the Point of View of Manifestations of Fatigue by the Determination of the "Factor of Elasticity." A. Buchwald. (Génie Civil, 1935, vol. 107, Aug. 24, pp. 185–186). The author criticises the usual method of determining the behaviour of bridges under dynamic loading by reference to the resistance offered by girders under alternating stresses, and advocates the use of a "factor of elasticity," the purpose of which is to allow for the possibility that the elasticity of the structure is not perfect. The author considers that for the materials of construction, particularly steels, used in bridge work, larger safety factors should be adopted than are now usual, because the stresses imposed by vibration and dynamic effects are greater than is revealed by fatigue tests in the laboratory.

Rear Axle Gears: Factors which Influence their Life. J. O. Almen and A. L. Boegehold. (American Society for Testing Materials, 1935, Preprint No. 27). The authors describe a series of endurance tests carried out on rear axle gears, which show that gear tooth failures may be plotted on the familiar S/N curve when the gear stresses are properly calculated. It is also shown that the Lewis formula for calculating gear tooth stress leads to serious errors. It is indicated that the chief benefit obtained from the use of alloy steels is the reduction in the amount of distortion, which results in lower stress concentration through more uniform tooth contact. The results of rotating-beam plain and notched-bar fatigue tests of several gear steels are also presented.

Spark Testing. (Iron Age, 1935, vol. 136, Sept. 26, pp. 32–37; Oct. 3, pp. 22–24). The spark-test characteristics of a number of the more important American S.A.E. steels are described and illustrated by spark-test photographs. The equipment required for carrying out the tests consists of a small high-speed electric grinder, coloured glasses for the operator, and an enclosed space to give semi-darkness. The spark streams of most plain carbon and alloy steels of the type used in automobile and engineering work are sufficiently distinctive to enable sorting to be carried out on that basis.

The Diffusion of Gases Through Metals. (Metallurgist, 1935, Oct. 25, pp. 68–71). This article constitutes a summary of a paper by C. J. Smithells and C. E. Ransley on the above subject (Proceedings of the Royal Society, May 1, 1935). These investigators measured the rates of diffusion of various gases through nickel, copper, molybdenum and iron tubes. The effects of temperature, pressure and surface condition were studied. In the

light of their results the authors consider that diffusion must be preceded by adsorption, and that both dissociation of the molecules and adsorption play a part in the phenomena observed.

The Use of Titanium in Alloying. (Metallurgist, 1935, June, pp. 47–48). Reference is made to the use of titanium as a deoxidising agent and age-hardening element, and to its affinity for carbon. To this latter property may be attributed the inhibition of weld-decay in chromium steels containing titanium. The use of titanium in non-ferrous alloys is also considered.

Zirconium and its Uses. L. Sanderson. (Sands, Clays and Minerals, 1935, vol. 2, June, pp. 47–50). Zirconium is a rare element which takes its name from the orthosilicate, zircon ($ZrSiO_4$). Baddeleyite, the natural oxide of zirconium, is more common than zircon, and is the principal source of the metal. The extraction of the pure metal is rather difficult, as the oxide is very stable. Perhaps the most important use for zirconium is in the alloy ferro-zirconium, manufactured by the reduction of a mixture of zirconium and iron ores in the electric furnace. It contains from 30 to 40 per cent. of zirconium, together with small amounts of carbon, aluminium and titanium. Ferro-zirconium alloys can be used for electric lamp filaments and spark points. The oxide zirconia is highly refractory, and is used for muffles, crucibles and furnace linings. Ferro-zirconium may be used as a scavenger in steel, but does not seem to have any advantages over ferro-titanium in this respect.

The Properties of Some Low-Nickel Steels Containing Manganese. R. H. Greaves. (Iron and Steel Institute, Sept., 1935: this Journal, pp. 99).

Aluminium in Modern Commercial Steels. H. W. McQuaid. (Iron Age, 1935, vol. 136, Oct. 10, pp. 20–27; Oct. 17, pp. 28–34, 84–88). The author follows some introductory remarks on the influence of aluminium on the grain-size of steel with a short sketch of the historical development of grain-size control, and then proceeds to discuss the results of some investigations on aluminium as a deoxidiser and grain-size control. He shows that aluminium not only allows the structure of the steel to be regulated, but that in this way the correct grain-size to give the best response to quenching and normalising and to confer the greatest toughness and hardenability (when carburised) may be established. Manganese affects the hardenability of quenched low- and medium-carbon steels; the influence of manganese, with or without aluminium additions, is represented in a series of diagrams. The effect of aluminium in refining the grain of steels is usually attributed to the submicroscopic particles of alumina dispersed throughout the molten steel. There are, the author points out, circumstances which suggest that this is not a sufficient explanation, and that other causes may be operative.

Thus it seems possible that it is the excess aluminium, over that required for deoxidation, which is responsible for its effect on the structure of the steel; this is borne out by the fact that both silicon and manganese, when used as deoxidisers, require the use of less aluminium. The author concludes with an examination of the effect of aluminium on the hardenability; in this connection he advances the theory that the aluminium influences the coalescence of the carbides precipitated during the austenite transformation.

The Development and Uses of Low Alloy High Tensile Steels. A. E. Gibson. (Journal of the American Welding Society, 1935, vol. 14, Sept., pp. 2–8). Attention is drawn to the economies made possible by the adoption of high-tensile low-alloy steels containing small proportions of copper, nickel, chromium, manganese, silicon and molybdenum, in place of the usual low-carbon steels. Special reference is made to their use in welded constructions.

The Development of High-Tensile Steels in Various Countries. J. Welter. (Ossature Métallique, 1934, vol. 3, pp. 573–586: Stahl und Eisen, 1935, vol. 55, July, pp. 736–737). The author reviews the high-tensile steels that have been developed in various countries; he gives a table showing the average compositions of a number of the more important steels.

A New High Speed Steel. F. Garratt. (Metal Progress, 1935, vol. 27, pp. 38–43). A high-speed molybdenum steel with properties comparable to those of the familiar tungsten steel containing 18 per cent. of tungsten, 4 per cent. of chromium, and 1 per cent. of vanadium, is described. The chemical analysis of a typical specimen of the new alloy is: Carbon 0·76, manganese 0·30, silicon 0·30, tungsten 1·60, molybdenum 7·90, chromium 3·85, and vanadium 1·10 per cent. The heat treatment is similar to that adopted for tungsten steels, and any tendency towards the development of a soft skin may be eliminated by using borax as a protective film; the articles should be dipped into a strong solution of borax before being transferred to the heat-treatment furnace. Molybdenum tool steel is claimed to be tougher and harder than the standard 18/4/1 alloy. It is marketed under the trade name "Motung."

New Steel Has High Machinability and Unusual Carburising Properties. (Steel, 1935, vol. 97, Aug. 19, p. 28). Sponsored by the Monarch Steel Corporation, Indianapolis, U.S.A., the new steel is a high, sulphur-high-manganese, low-carbon steel, known as Speed Case. It is said to be highly machinable and readily carburised; for the cold-drawn state in which it is usually marketed a tensile strength of 85,000 lb. per sq. in., an elongation on 2 in. of 26 per cent., and an Izod value of 60–80 ft.-lb. are claimed.

Bright and Free-Cutting Steels. R. T. Rolfe. (Iron and Steel Industry, 1935, vol. 8, May, pp. 315–318; June, pp. 341–345). The use of bright bars and strip carries with it many advantages, notably a reduction in machining costs owing to the absence of scale. The cold-work which the bright steel undergoes in the process of cold-drawing or rolling usually improves the ultimate strength but may seriously diminish the impact figure. Cold-work tends to divorce the pearlite, and when such material is subjected to improper heat treatment the cementite coalesces and a low impact value results. Hard-drawn material should be normalised in order to obtain adequate shock resistance. Free-cutting steels usually have comparatively high phosphorus or high sulphur and manganese contents. By the use of these steels a considerable reduction in machining costs may be obtained; if, however, the free-cutting qualities are gained by introducing phosphorus the resultant material may be completely lacking in resistance to shock; this is especially the case if the steel is also finished by cold-drawing. Many failures may be attributed to the use of bright-drawn free-cutting steels which have not received the proper normalising treatment. High phosphorus in a steel is likely to give rise to a weak banded structure, together with pearlite coalescence. Normalising, followed by oil-quenching, may be necessary to correct this. Free-cutting properties in a steel may be obtained by using increased sulphur and manganese contents, and such steels generally have a more satisfactory impact resistance.

Free-Machining Selenium-Containing Alloy Steels Covered by Patents. (Steel, 1935, vol. 97, Aug. 12, pp. 39–40). Patents covering the use of selenium in alloy steels have been granted to the Carpenter Steel Co., Reading, Pa., U.S.A. It is claimed that selenium imparts free-machining properties, and that it is superior to sulphur in this respect; however, it is admitted that the high cost constitutes a disadvantage, and, so far, the experiments have been confined to electric steels.

Modern Alloys in Marine Engineering and Shipbuilding. J. W. Donaldson. (Metal Treatment, 1935, vol. 1, Summer Number, pp. 59–62). The author discusses the uses and applications of special steels, cast irons and non-ferrous alloys for the construction of ships' hulls, marine boilers, steam reciprocating engines, turbines, oil engines and auxiliary machinery.

Alloy Steels for Aircraft Construction. W. H. Hatfield. (Metallurgia, 1935, vol. 12, June, pp. 45–48). The author commences with a reference to the importance of proof stress, fatigue stress and ductility, from the point of view of safety, and proceeds to survey the range of materials available for aircraft construction. Tables giving the strength/weight ratios, and, in the case of ferrous alloys, the physical properties, heat treatment, and chemical

composition, are reproduced, and the applications of a number of these steels are shortly considered.

Nickel Alloys in Automobile Construction. J. A. Rabbitt. (Japan Nickel Review, 1935, vol. 3, July, pp. 292–362). A survey is given of the many uses to which nickel steels are put in the fabrication of motor vehicles. The examples are mainly drawn from English and American practice, and include steels employed for engine, transmission, chassis and bodywork details. In each case the composition, mechanical properties and service duty are cited. The composition and properties of nickel alloy cast irons are quoted, and examples of the application of this material to cylinder heads, crankshaft and camshaft castings, and brake drums are furnished; the melting and heat treatment practice is also given. Brief attention is devoted to certain non-ferrous nickel alloys. The data are summarised in tables dealing with the specifications, performance requirements, mechanical properties and heat treatments of nickel-alloy steels and cast irons.

Nickel Alloys in Three-Wheel Delivery Cars. H. Okuda. (Japan Nickel Review, 1935, vol. 3, July, pp. 398–404). The importance of nickel alloys, and especially nickel cast iron, in the successful development of three-wheeled motor vehicles, is indicated.

Nickel Alloys used in General Motors' Products in Japan. S. E. Dithmer. (Japan Nickel Review, 1935, vol. 3, July, pp. 363–367). The principal parts into which nickel enters as an alloying element, in the production of motor cars by the General Motors Corporation, are briefly enumerated.

Alloy Sheet Steel (Copper-Nickel-Molybdenum). H. L. Miller. (Metal Progress, 1935, vol. 28, July, pp. 28–31, 70). The properties and applications of a copper-nickel-molybdenum steel sponsored by the Republic Steel Corporation are dealt with. This steel is marketed in strip or sheet form, and has the following analysis: Manganese 0·50 to 1·00, copper 0·50 to 1·50, nickel 0·40 to 0·80 per cent., and molybdenum 0·20 per cent. (max.). Two grades are made, one containing 0·12 per cent. of carbon, and the other 0·30 per cent. In the form of plate the low-carbon material gives an ultimate strength of 80,000 lb. per sq. in., a yield point of 70,000 lb. per sq. in., and an elongation on 2 in. of 30 per cent.; the medium-carbon steel has an ultimate strength of 90,000 lb. per sq. in., a yield point of 70,000 lb. per sq. in., and an elongation of 20 per cent. on 2 in. The nickel and copper are added to strengthen the ferrite, and the molybdenum to refine the grain and retain the copper in solution. It is possible to work, weld, and fabricate nickel-copper-chromium steel sheet quite readily, and its good resistance to corrosion is an additional advantage.

Manganese Steel in Crushing Machines. W. B. Pickering. (Sands, Clays and Minerals, 1935, vol. 2, June, pp. 73–80). Crushing machines in which manganese steel is employed for parts required to withstand severe abrasion are described. These machines include, single toggle breakers, roll crushers, gyratory crushers and disc crushers.

Copper Steel Castings. (Metallurgist, 1935, Aug., pp. 58–60). A summary in English of a paper by E. Söhnchen and E. Piwowarsky which appeared in Die Giesserei, 1935, vol. 22, Mar. 1, pp. 96–100, is provided. (*See* Journ. I. and S.I., 1935, No. I. p. 492.)

Steel Castings in Mining Operations. J. A. de Bondy. (Canadian Mining and Metallurgical Bulletin, 1935, No. 281, Sept., pp. 453–457). Some notes on the materials used for steel castings in mining practice are provided, manganese steel being the material chiefly dealt with.

The Influence of the Material on the Life and Economy of Mine-Filling Blast Pipes. K. Bax. (Glückauf, 1935, vol. 71, Sept. 21, pp. 913–917). Ludwig and Pande, in two recent publications, have come to the conclusion that for mine-filling blast pipes the most suitable and economical material is ordinary weldless steel tube with a tensile strength of 55–65 kg. per sq. mm., and that all other kinds of pipe material, particularly hardened steel, increase the cost of mine-filling by this method excessively. The present author points out that this conclusion is unexpected, having regard to the superiority of hardened parts for wear resistance in most other cases. He submits Pande's test results to critical examination; it is clear that, whilst hardened steel tubing was very much more wear-resistant on the testing machine, yet, nevertheless, in tests underground the use of hardened tubes appeared uneconomic; certainly the life of the hardened tubes exceeded that of the unhardened ones, but, according to Pande, this was insufficient to balance their higher first cost and, so resulted in an increase in the cost of gobbing. The author then gives consideration to the possibilities of error in the results of Ludwig and Pande, and concludes that their work does not provide a final decision as to the economy of the various materials for mine-filling blast pipes.

Wire Ropes. (Colliery Guardian, 1935, vol. 151, Aug. 2, pp. 197–198; Aug. 9, pp. 242–244; Aug. 23, pp. 332–333). These articles contain information derived from the Fifth Report of the Wire Ropes Research Committee of the Institution of Mechanical Engineers. Among the subjects dealt with are: Rope speed; lubrication; flattened strand ropes; the relation between pulley, wire and rope diameters; and the effect of the tensile strength of the wire on the life of the ropes. The results of the tests which were carried out are reproduced. An investiga-

tion on the effect of lubricants revealed that dry ropes generally possess a superior endurance. Data on the coefficient of friction of lubricated and non-lubricated ropes are also given.

The Deterioration of Haulage Ropes in Service. S. M. Dixon and M. A. Hogan. (Mines Department, Safety in Mines Research Board, Paper No. 92, 1935). The authors report the results of an investigation of the causes of failure of haulage ropes (used for hauling men), and indicate the relative importance of the various factors which caused the deterioration of the ropes. The circumstances attending the breakage of the ropes are briefly considered, special attention being called to those which were of an unusual character. Precautions are suggested for the prevention of the occurrence of similar failures, and the practical conclusions arrived at during the investigation are set out in such a manner as to be of assistance to colliery managers in the selection, installation and maintenance of haulage ropes and in the capping of such ropes. The rope failures are dealt with in detail, grouped according to whether the failure of the rope, as judged by the tested samples and the results of mechanical tests, was found to be due to wear, corrosion or capel failure.

Economical Selection of Sucker Rods. B. B. Wescott and C. Norman Bowers. (Transactions of the American Institute of Mining and Metallurgical Engineers, 1935, vol. 114 : Metals Technology, 1935, vol. 2, June). This article refers to the selection of the materials available for the pumping rods of oil wells. The types of oil-field met with may be divided into (1) non-corrosive, (2) corrosive, non-sulphide, (3) corrosive, sulphide. Satisfactory sucker-rods for non-corrosive wells may be readily obtained, but the same is not equally true of the other types of well. The authors supply figures for the relative cost, chemical composition, physical properties, and corrosion-fatigue limits (when subjected to the action of hydrogen sulphide), of a number of typical sucker-rod materials. Data giving the degree of embrittlement under hydrogen sulphide are furnished, and stress-concentration factors for notched specimens are also reproduced. On the basis of their results the authors suggest that the pump rod material should have a yield point of 50,000–60,000 lb. per sq. in. in order to permit the redistribution of local stress where corrosion pits occur, and that the carbon content should be below 0·10 per cent., so as to secure maximum toughness. An empirical formula has been deduced relating the yield point, sulphide corrosion-fatigue limit, and operating conditions, and indices have been calculated for each material. Finally, the sucker-rod materials are classified according to the service conditions to be met.

Low Alloy Steels for Oil Refinery Service. R. L. Wilson. (Metal Progress, 1935, vol. 28, Sept., pp. 29–33). Steels containing

from 1 to 6 per cent. of chromium are largely employed for oil refinery work. With up to about 2 per cent. of chromium the steel exhibits a relatively greater resistance to corrosion by crude oil, compared with the amount of chromium present, than steels of higher chromium content. The resistance of chromium steels is greatly improved by the addition of silicon ; a steel containing 2·75 per cent. each of silicon and chromium is the equivalent of a 5 per cent. chromium 1·25 per cent. silicon steel in this respect. If temper-brittleness is to be avoided a molybdenum-chromium steel with 0·5 per cent of molybdenum should be employed ; this element also improves the creep strength. Of the steels suitable for use in oil refinery work the following low-chromium steels have been found useful : A non-air-hardening, high creep-strength steel containing carbon 0·15, silicon 0·75, chromium 1·25 and molybdenum 0·50 per cent. ; and a steel with a similar analysis but having double the quantity of chromium.

Materials Used in Oil-Refinery Pumps. A. E. Harnsberger. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 631). The materials are considered under the following headings : General service pumps ; pumps used on distillation and cracking equipment, handling oils at up to 200° F. ; pumps for service at up to 900° F. ; pumps used on gas absorption plants ; sludge acid pumps ; pumps for miscellaneous processes. In each case the materials (ferrous and non-ferrous) used for the various parts of reciprocating, rotary and centrifugal pumps, are described.

Some Mechanical and Metallurgical Aspects of Present-Day Oil-Production Equipment. A. G. Zima. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 633). The steels used for items such as rock drill and reamer cutters, rotary machine gears, bodies for drills, reamers, &c., are discussed, with reference to their suitability for the tasks they are required to perform.

Note on the Use of Link Pins of 12–14 per cent. Manganese Steel in the Bucket Chains of Dredgers. A. Caulle. (Arts et Métiers, 1935, vol. 87, Aug., pp. 166–175). The author gives an account of researches made to discover the cause of the premature breakage of link pins, made of 12–14 per cent. manganese steel, in the bucket chains of certain dredgers, at a time when the degree of wear was still insufficient to require their replacement. He concludes that the principal cause lay in the fact that pins which, owing to the shape of their heads, are capable of rotating, are submitted to alternating bending, which flexure is sufficient to deepen the minute cracks forming on the cold-worked surface of the pins. He recommends that such designs of pins should be prohibited ; when using 12 per cent. manganese steel, care should be taken that the wearing parts are subjected to pressure

only, without alternating bending, as the latter causes the deterioration of the superficial cold-worked surface—a characteristic of this type of steel.

Alloy Steels for Boiler Drums and Superheaters. G. K. Herzog. (Blast Furnace and Steel Plant, 1935, vol. 23, Aug., pp. 554–556, 574). The author remarks that the high temperatures which now obtain in boiler practice necessitate attention being given to the strength at elevated temperatures, especially the creep strength, of carbon and alloy steels. He considers the properties of the latter in more detail, with respect to the alloying elements employed.

Developing Alloy Steels for High-Pressure Boilers and Pressure Vessels. L. P. McAllister. (Steel, 1935, vol. 97, Aug. 5, pp. 30–33, 48). It is shown that the demands of the welder, coupled with an increase in boiler pressures, have directed attention towards the alloy steels. The advantages of low-alloy steels containing nickel, manganese, vanadium and molybdenum are stressed, and examples of their use are quoted.

Modern Materials for High-Pressure Boiler and Pressure Vessels. L. P. McAllister. (National Boiler and Pressure Vessel Inspectors, Chicago, May 16, 1935: Iron Age, 1935, vol. 135, June 6, pp. 24–26, 108, 110, 112). Designers tend more and more to favour the adoption of high-strength materials for pressure vessels; this involves an improvement in tensile strength from about 55,000 lb. per sq. in. to 75,000 lb. per sq. in., and, generally, a corresponding increase in the carbon content of the steel. There is a limit to the extent to which the strength of plain carbon steel may thus be raised, and in order to retain the advantages associated with the use of high-tensile steel recourse must be had to alloys. Alloy steels enable superior high-temperature creep, fatigue, impact and corrosion-resistance properties to be obtained, in addition to a reduction in the weight of the finished article and, possibly, in the cost of fabrication. Such steels contain nickel, chromium, silicon, molybdenum and vanadium, quite small amounts of these elements serving markedly to improve the physical properties of a normal low-carbon steel. Thus, a 2 per cent. nickel steel (75,000 lb. per sq. in. tensile strength) has been used to advantage in Canadian locomotive construction. Manganese-vanadium and carbon-molybdenum steels are also useful low-alloy materials.

Cause of Fissures in Riveted Boiler Drums. R. Rist. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, June 29, pp. 812–813). The author has examined a number of cases of fissuring in boilers in an attempt to diagnose their cause. Features for observation were (*a*) the position of the fissures, (*b*) the form of the fractured surface, and (*c*) the path of the fracture through the metallographic constituents of the metal. No regularity in the positions at which fracture occurred was observed; fatigue alone, therefore, could not be the cause of all of them, as in that case

they would always have been found in similar positions, where the stresses were greatest. In general, the types of fracture could be divided into two groups: (1) Fractures with a granular structure in riveted joints and shell plates, and (2) fractures with a smooth grainless structure of the fatigue type in bottom flanges; at the same time, one of the fractures in a riveted joint showed clear evidence that it was a fatigue fracture. The first type of fracture was usually ramified and ran through the grain boundaries, whereas the second type usually followed a straight path. The starting point for the first type of fracture was frequently a cavity in the interior of the metal. The author concludes that in the process of fracturing at riveted joints and in shell plates, the effect of intercrystalline corrosion is more important that that of fatigue.

Relation of Chemical Composition to the Rail Fissure Problem. E. Cook. (Transactions of the American Society for Metals, 1935, vol. 23, June, pp. 545–555). The author discusses certain aspects of the problem of rail fissuration, with special reference to the influence of composition and low temperatures.

Contribution to Knowledge of the Changes with Time of the Mechanical Properties of Newly-Rolled Rails, Particularly of Basic Bessemer Steel. F. Körber and J. Mehovar. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1935, vol. 17, No. 7, pp. 89–105). During the determination of the mechanical properties of newly-rolled and straightened open-hearth steel rails, it was observed that the properties varied considerably according to whether the tensile specimens were pulled immediately after rolling or after an interval; storage caused a considerable rise in the elongation and reduction of area. This phenomenon was examined further; tests were made on basic Bessemer silicon-bearing steel rails (S.49), with a tensile strength of 70–77 kg. per sq. mm., and, for comparison, on open-hearth steel rails of the same silicon content (about 0·85 per cent.), tensile strength and profile. The investigation was directed primarily to the determination of the changes in the yield point, tensile strength, elongation and reduction of area. In addition to observing the magnitude, the changes with time and the possibilities of influencing the mechanical properties, the authors have also sought for an explanation of the facts noted.

In tests regarding the influence of the period of storage and of tempering (which had a parallel effect) on complete rail sections and on finish-machined specimens (of both unstraightened and straightened rails of basic Bessemer and, to some extent, open-hearth rails), it was found that: (*a*) Annealing the specimens at 200° to 600° C. resulted in an immediate improvement of the properties; (*b*) on storing the test-pieces, the improvement set in after days or weeks; and (*c*) on storing the entire rails, the properties required months to become stable. The attainment of the

best values by storage proceeded more slowly in open-hearth than in basic Bessemer rails.

In contradiction to the opinion expressed in the literature, the rolling and cooling stresses, likewise the cold-straightening stresses, had very little influence on the phenomena ; on the contrary, it is conjectured that the cause of the changes of properties on storage or on tempering lies in the equalisation of structural stresses. This supposition is supported by the fact that, except for changes observed after a hot-deformation (-rolling) treatment, the same time- and temperature-dependent improvements were found after a recrystallising anneal ; further, the more complete the transformation, the greater was the rise in the elongation and reduction-of-area values. Direct detection of the microscopic stresses with the aid of X-rays was not possible. The results recorded here differ radically from those yielded by treatments giving rise to precipitation phenomena ; consequently the causes must be of a different nature. No explanation could be found for the frequently observed difference in the behaviour of basic Bessemer, open-hearth and alloyed open-hearth steels.

The importance of the observations here recorded from the point of view of acceptance tests, for instance, is well demonstrated by a collation of the results, which shows that if the tensile specimens of newly-rolled rails are given a short tempering treatment, whilst the yield point and tensile strength remain practically unaltered, the elongation will increase by about 30 per cent., and the reduction of area by even as much as 100 per cent.

Considerations on the Sudden Fracture of an Electric Power Line Pylon. J. Merklen and E. Vallot. (Génie Civil, 1935, vol. 106, June 22, pp. 612–614). The authors present the results of their investigations of the sudden fracture of a grid pylon during a heavy storm. Macro- and micrographs of the angle-iron which failed were prepared, and tests to determine its toughness were made on specimens cut from it ; it was found that the steel was not sound, exhibited heterogeneity and was very brittle just at that part of the section which should possess maximum strength. The authors assert that macro- and micro-examination correlated with brittleness tests would have revealed the defectiveness of the material, and they discuss the inclusion of such tests in specifications.

SPECIFICATIONS.

British Standard Specification for Cast Iron Spigot and Socket Soil, Waste, Ventilating and Heavy Rainwater Pipes. (British Standards Institution, No. 416–1935).

British Standard Specification for the Dimensions of Grinding Wheels and Method of Attachment. (British Standards Institution, No. 620–1935).

British Standard Specification for Wire Ropes of Special Construction for Engineering Purposes, inclusive of Cranes, Lifts and Excavators. (British Standards Institution, No. 621–1935).

MAGNETIC AND ELECTRICAL PROPERTIES OF IRON AND STEEL.

A New Double-Yoke Magnet Steel Tester. F. Stäblein and R. Steinitz. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, June, pp. 549–554: Technische Mitteilungen Krupp, 1935, vol. 3, No. 3, pp. 129–135). The new instrument for testing magnet steels consists of two E-shaped half-yokes of soft iron turned towards each other so as to form a magnetic circuit; the three cross members are separated by air-gaps. The specimen is placed in one of the outer air-gaps, the width of which can be varied; the middle one is used for measuring the *B–H* value of the specimen (by means of a moving coil, for instance). The outer yoke arms carry the exciter coils, which are connected in series in such a way that without a specimen in position the lines of force form a circuit which, on account of the symmetry, does not pass through the middle leg. By placing the specimen in one of the outer air-gaps the symmetry is destroyed, and the flow of force across the middle leg gives a measure of the *B–H* value of the specimen.

The Specification of Magnetic Qualities. L. G. A. Sims. (British Association, Sept. 9, 1935: Engineering, 1935, vol. 140, Sept. 13, pp. 290–291). The author discusses the methods of measuring magnetic qualities, particularly permeability.

Rolled and Recrystallised Textures of Iron-Nickel Alloys in Relation to the Magnetic Properties. F. Pawlek. (Zeitschrift für Metallkunde, 1935, vol. 27, July, pp. 160–165). The production and removal of the rolled and recrystallised textures of iron-nickel alloys are discussed in their individual phases. The marked influence of the various textures on the properties of the material is shown, and the commercial importance of the magnetic properties obtained is indicated.

The Rolled and Recrystallised Textures of Nickel-Iron. W. G. Burgers and J. L. Snoek. (Zeitschrift für Metallkunde, 1935, vol. 27, July, pp. 158–160). By means of X-rays, using molybdenum K_α radiation, the rolled and recrystallised textures of a 50 per cent. nickel-iron alloy have been determined.

The Magnetic Properties of Sheet Iron Produced Electrolytically (Dependence on Grain-Size, Sheet Thickness and

Arsenic Addition). O. Dahl, F. Pawlek and J. Pfaffenberger. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Aug., pp. 103–112). On the basis of tests carried out on electrolytic sheet iron, the authors discuss the magnetic behaviour and properties of the material under normal conditions and the dependence of the watt losses on the sheet thickness and grain-size. They next describe their attempts to reduce the electrical losses due to eddy currents and magnetic fatigue by the introduction of an alloying element capable of forming a solid solution ; nickel, copper, zinc, antimony and tin were co-deposited during the electrolytic production of the sheet, but the results were not promising ; by a diffusion process, arsenic was introduced into the sheet, and with 0·5 to 1 per cent. the electrical losses were greatly reduced.

Effect of Carbon, Oxygen and Grain Size on the Magnetic Properties of Iron-Silicon Alloys. T. D. Yensen and N. A. Ziegler. (American Society for Metals, Oct., 1935, Preprint No. 13). This paper is a continuation of a previous one (Transactions of the American Society for Metals, 1935, vol. 23, June, p. 556), and gives numerical relationships between the carbon and oxygen contents and the grain-size and the magnetic properties of iron-silicon alloys. Oxygen is without appreciable effect on the low-density properties, but affects the high-density permeability considerably. The effect of carbon is generally less and the effect of grain-size is generally greater (for low carbon contents) than previous results indicated (Transactions of the American Institute of Electrical Engineers, 1924, vol. 43, p. 145) ; but the effect of the various forms of carbon (dissolved or colloidal Fe_3C, free Fe_3C, pearlite, graphite) as shown in the 1924 paper has been confirmed.

Nickel-Aluminium Magnet Steel. (Metallurgist, 1935, June, pp. 39–41). A review of some work recently carried out by Messkin and Somin on the effect of heat treatment on the magnetic properties of nickel-aluminium magnet steel is given. (*See* Journ. I. and S.I., 1935, No. I. p. 497).

The Recent Development of Materials for Permanent Magnets. F. Pölzguter. (Stahl und Eisen, 1935, vol. 55, Aug. 8, pp. 853–859). The author reviews the development of the materials for permanent magnets, from tungsten steels, through cobalt steels, up to the latest aluminium-nickel and cobalt-titanium alloys ; particular attention is given to the modern steels whose superior properties are the result of precipitation hardening. He then deals with the principal steels so far generally used in practice, their composition, heat treatment, production and properties, and concludes by indicating the shapes of magnets required by the special properties of the new alloys, and some new fields of application.

Recent Progress in Alloys for Permanent Magnets.

(Metallurgist, 1935, Oct. 25, pp. 77–79). This is an English summary of the preceding article by F. Pölzguter.

Colbalt Magnet Steels. L. Sanderson. (Heat Treating and Forging, 1935, vol. 21, Aug., pp. 367–368). Chief attention is paid to the heat treatment of cobalt magnet steels, and this is followed by a short note on the magnetising of these steels and their preparation in the form of finished magnets.

New Materials for Permanent Magnets. A. Kussmann. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Sept. 28, pp. 1171–1173). After touching on the relation between the design and the magnetic behaviour of a magnet, the author discusses the properties of the latest materials used for permanent magnets, such as tungsten steel, cobalt steel, and alloys of iron with cobalt-molybdenum, nickel-copper, nickel-aluminium and cobalt-titanium. He also mentions sintered and pressed magnets —the so-called oxide magnets—made by sintering together certain ferromagnetic oxides, in particular, cobalt ferrite ($CoFe_2O_4$) and iron oxide ($FeFe_2O_4$), and concludes with a reference to the platinum-iron alloys (about 78 per cent. platinum).

X-Ray Investigations on α-Iron Magnet Alloys. R. Glocker, H. Pfister and P. Wiest. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, June, pp. 561–563). In experiments on permanent magnet steels containing (*a*) aluminium 12 and nickel 30 per cent., and (*b*) aluminium 10, nickel 22½ and cobalt 13 per cent., changes in the radiographs occasioned by various heat treatments were observed and correlated with variations in the coercive force. When the coercive force was at a maximum there was no broadening of the lines, neither was there any change in the lattice constants. Only after very prolonged tempering at high temperatures, causing the coercive force to become very small, was any change of lattice constant noticeable. At still higher temperatures the lines of the precipitated phase appeared ; this belonged to the α-iron lattice type. A clearer idea of the concept " precipitation " is presented on the basis of considerations of the physics of the lattice. Precipitation exists only when the atoms of the precipitating phase have left the combined lattice. In this sense, in the steels investigated, as in the hardening of duralumin, when the coercive force is at a maximum precipitation has not occurred. There still remains the problem of the optimum particle size of the precipitated phase from the point of view of the magnetic properties.

Electrical Resistance Metals. A. Schulze. (Giesserei, 1935, vol. 22, June 21, pp. 312–314). The compositions and physical properties of a number of resistance alloys (non-ferrous and ferrous), both for precision resistances and for industrial purposes, are tabulated.

CONSTITUTION, STRUCTURE, CRYSTALLOGRAPHY.

Research and the Iron and Steel Industry. II.—Constitution, Grain Size, Alloy Steels, Cast Irons. J. W. Donaldson. (Iron and Steel Industry, 1935, vol. 8, Aug., pp. 417–420). This article constitutes a review of the more important researches recently undertaken in the fields of alloy constitution, grain-size, alloy steels, and cast irons.

The Metallurgical Inspection of Engineering Materials. V. E. Green. (Proceedings of the Staffordshire Iron and Steel Institute, 1933–1934, vol. 49, pp. 66–70). The reasons for inspection, methods of inspection, place of inspection and qualifications and duties of an inspector are successively dealt with.

Ten Years of Metallurgy and Science of Ferrous Materials. R. Walzel, R. Mitsche and H. Pessl. (Berg- und Hüttenmännisches Jahrbuch, 1935, vol. 83, No. 2, pp. 61–76). The authors present a review of advances made in the metallurgy of iron and in the development of its alloys and their uses.

Definitions of Technical Terms. K. Daeves. (Stahl und Eisen, 1935, vol. 55, Sept. 5, p. 962). The definitions of terms given were drawn up by the author at the request of the Werkstoffausschuss des Vereins deutscher Eisenhüttenleute. Three new definitions are given, and eight others, originally proposed in 1929, are repeated as being now completely acceptable. The words defined are:

Altern	*Glühen*
Reckaltern (Stauchaltern)	*Normalglühen*
Aushärtung	*Weichglühen (auf körnigen Zementit glühen)*
Wärmebehandlung	*Anlassen*
Abschrecken	*Vergüten*
Härten	

Oxalic Acid as an Electrolytic Etching Reagent for Stainless Steels. G. A. Ellinger. (American Society for Metals, Oct., 1935, Preprint No. 2). A useful method for the etching of corrosion-resistant alloys is described. The solution generally used contains 10 grm. of oxalic acid dissolved in 100 ml. of distilled water, which may be diluted if desired for specific purposes. The polished specimen is made the anode and a piece of platinum gauze the cathode, the two being about 1 in. apart. Current from a 6-v. accumulator or from four dry batteries connected in series is passed through the cell. The method reveals both the carbides and grain boundaries in most alloys of the 18/8 type, after suitable etching periods. Alloys containing selenium are not completely etched.

On the Preparation of Iron and Steel Specimens for Microscopic Investigations. F. F. Lucas. (American Society for Metals,

Oct., 1935, Preprint No. 1). The author describes methods and materials for the preparation of iron and steel specimens. A flotation apparatus for the preparation of abrasives is described, and the particle size of magnesium oxide abrasive prepared by this method is given.

On the Nature of Polished Layer of Metals. M. Miwa. (Science Reports of Tôhoku Imperial University, 1935, vol. 24, July, pp. 222–239). The well-polished surfaces of various metals were examined by cathode ray diffraction, and it was shown that these surfaces may, except in a few cases, be considered to be amorphous. Metal surfaces in various stages of polishing were also studied, and it was found that as a result of continued polishing the diffraction lines became blurred. It is suggested that in the case of cadmium, tin and antimony (which metals do not appear to give an amorphous layer), the surface of the metal consists of tiny pseudo-crystals, scarcely distinguishable from particles in the liquid state ; further, it is supposed that ultimately there is no sharp distinction between the amorphous and crystalline states, and that a polished surface is of this nature.

Principles of the Technique of Photography for Use in Metallography. R. Servant. (Métaux, 1935, vol. 10, May, pp. 115–130). The author discusses the objectives and photographic plates and papers used in metallography from the theoretical and practical standpoints ; he also presents some practical observations regarding the making of photomicrographs.

The Interpretation of a Photomicrograph. J. W. Jones. (Proceedings of the Staffordshire Iron and Steel Institute, 1933–1934, vol. 49, pp. 48–61). The author considers typical microstructures and their interpretation.

A Small X-Ray Apparatus for Microstructural Investigations. W. E. Schmidt. (Zeitschrift für technische Physik, 1935, vol. 16, pp. 115–119). A small X-ray apparatus, consuming about 450 watts, for investigating the microstructures of materials is described.

The Researches of F. Regler on the Testing of Materials by Means of X-Rays. F. Wever and H. Möller. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, July, pp. 47–55). The authors present a review and criticism of the publications of F. Regler on his investigations of fine structure and, in particular, stress measurement by means of X-rays. Regler's reply to the criticism is appended.

Interior Defects Revealed by X-Ray. F. K. Ziegler. (Metal Progress, 1935, vol. 27, June, pp. 45–48). Examples illustrating the value of X-ray examination in detecting otherwise invisible flaws in castings are furnished, together with radiographs of typical defects.

Grain-Growth in Carbonyl Iron and the Production of Single Crystals of Iron. W. Tangerding. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Aug., pp. 113–114). Oxidising or reducing annealing of carbonyl iron caused an unusual increase in the grain-size. By means of alternating oxidising and reducing annealings single crystals of iron could be produced with ease. The veining of the ferrite in carbonyl iron is largely dependent on the rate of cooling. In comparison with specimens free from veining, the coercive force is raised decidedly by veining. This supports the view that the veining constituent consists of a layer with a distorted α-lattice which has absorbed very small quantities of impurities.

Crystal Growth and Crystal Nucleus Formation. J. N. Stranski and R. Kaischew. (Physikalische Zeitschrift, 1935, vol. 36, pp. 393–402). The authors present a new theory on the growth of crystals and nucleus formation, which they have developed by means of a simplified crystal model.

Hardening Characteristics of One per cent. Carbon Tool Steels. T. G. Digges and L. Jordan. (American Society for Testing Materials, Oct., 1935, Preprint No. 23). The authors have studied the hardening characteristics of two commercial 1 per cent. carbon tool steels. Specimens of each steel were prepared with three different initial structures, namely, spheroidised cementite, coarse pearlite, and sorbite, and a study was made of the effects of these initial structures on the austenitic grain-size and grain-growth characteristics and on the critical cooling rates. The data obtained made possible a direct comparison of the relations between the austenitic grain-size and critical cooling rate of the two steels over a range of quenching temperatures from 775° to 970° C.

Relation of Hot Working to McQuaid-Ehn Grain-Size. H. A. Grove. (American Society for Metals, Oct., 1935, Preprint No. 14). A brief study is made of the effect of the working temperature on the McQuaid-Ehn grain-size of a number of steels. Some of the irregularities in grain-size that can be produced by working at different temperatures are touched upon, and examples are given of duplexing developed in fine-grained steels by rolling and forging.

Effect of Grain-Size Upon Some Physical Properties of Medium Carbon Steels. F. G. Sefing and K. J. Trigger. (Transactions of the American Society for Metals, 1935, vol. 23, Sept., pp. 782–796). The authors show that grain-size has a profound influence on cracking and distortion in medium-carbon steels.

The Crystal Structure and the Composition of the Trigonal Chromium and Manganese Carbides. A. Westgren. (Jernkontorets Annaler, 1935, vol. 119, No. 6, pp. 231–240). In chromium steels with comparatively high chromium and carbon contents the carbide phase consists of a trigonal chromium carbide in which

some of the chromium is replaced by iron. This carbide is also present in high-carbon ferro-chromium, and an analogous manganese carbide is the main constituent of high-carbon ferro-manganese. These carbides are usually considered to have the formulæ Cr_7C_3 and Mn_7C_3. The author describes an attempt to determine the structure of the chromium carbide in order to test the validity of the formula. For this purpose the data of X-ray analysis have been drawn upon, and a tentative structure has been suggested. Although the structure could not be completely deduced, the approximate solution left no doubt that the formula assigned to the carbide (Cr_7C_3) is the correct one. By analogy it is deduced that manganese carbide is truly represented by Mn_7C_3.

The Components of Steel. K. R. Van Horn. (Metal Progress, 1935, vol. 28, Aug., pp. 22–27, 68). In this article the nature of the constituents present in steels—namely austenite, martensite, pearlite and sorbite—and their modes of formation are considered. It is explained that recent investigations indicate that austenite is an interstitial solid solution of carbon in γ-iron which is converted to lamellar pearlite at temperatures of about 400° to 700° C. Martensite appears to be an interstitial solid solution of carbon in tetragonal (body-centred) iron, and is not an intermediate product in the austenite–pearlite transformation. When martensite is aged or tempered iron carbide is precipitated, leaving ferrite. The carbide particle size is related to the quenching rate, and when the structure so produced is comparatively coarse it is termed sorbite.

On Naming the Aggregate Constituents in Steel. J. R. Vilella, G. E. Guellich and E. C. Bain. (American Society for Metals, Oct., 1935, Preprint No. 3: Metal Progress, 1935, vol. 28, July, pp. 28–33). A number of micrographs are used to illustrate and classify, from the structural standpoint, the modes of transformation of austenite and the tempering of martensite in carbon steel. These structural types are considered genetically, and a simple basis of nomenclature is set forth. The authors propose that the term " pearlite " be applied to all the lamellar (ferrite and carbide layer) structures formed directly by nodular growth from nuclei in the austenite at comparatively high sub-critical temperatures, and that the word " sorbite " be applied to tempered martensite structures. If the word " troostite " is to be retained it should preferably be applied in such a way as to introduce no ambiguities or confusion; at present a suitable category for troostite is not apparent, though some possibilities not wholly logical are mentioned. The term " spheroidite " is suggested to designate the more coarsely spheroidised carbide structures.

The Effect of Deoxidation on the Rate of Formation of Ferrite in Commercial Steels. D. L. McBride, C. H. Herty, jun., and

R. F. Mehl. (American Society for Metals, Oct., 1935, Preprint No. 28). The rate of precipitation of ferrite from austenite at constant temperatures between A_3 and A_1 was studied in a series of nine commercial steels, including silicon-killed and aluminium-killed steels. The isothermal decomposition curves obtained can be described by the equation for a first-order chemical reaction, and the reaction rate in this equation can be taken as a measure of the reactivity of the steel. A steel treated to give different austenite grain-sizes gives different reaction rate constants for one temperature of precipitation; when, however, these are divided by the grain surface area they become equal. The rate of ferrite precipitation in aluminium-killed steels is greater than that in silicon-killed steels at low degrees of undercooling, but less at high degrees of undercooling, thus showing a smaller dependence on temperature. The amount of ferrite formed on normalising, compared with that formed on annealing, was found to be dependent only on the mass of the specimen, which determines the cooling rate, and the austenitic grain-size, increasing linearly with increasing austenite grain surface area per unit volume of specimen. (*See* Journ. I. and S.I. 1935, No. I. p. 503).

The Median Serration in Martensite Needles. E. Scheil. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, July, p. 61). A rib-like streak is often observed down the middle of martensite needles; it occurs in both low- and high-carbon steels, provided that considerable quantities of austenite exist in the structure near the martensite. This median serration is absent from deformation twins, which in other respects are very similar to martensite needles. During the tempering of martensite needles, cementite formation takes place preferably at the serrations. It was simple to explain the rib as being due to the different orientations of the two parts of the needle separated by it, but tests on the deformation of martensite needles shows them individually to be uniformly orientated, hence this explanation breaks down. No other suggestion has been proposed.

X-Ray Investigation of Austenite and Martensite in some Special Steels. Z. Nishiyama. (Science Reports of the Tôhoku Imperial University, 1935, vol. 24, May, pp. 128–140). Special steels containing chromium, manganese, or nickel were quenched from temperatures within the γ region and examined by X-rays. Austenite was found to have a face-centred cubic lattice and martensite a body-centred tetragonal lattice, as in the case of carbon steel. The martensite derived from austenitic nickel steel by cooling in liquid nitrogen also has a body-centred tetragonal lattice, but the martensitic structure found in quenched and tempered high-carbon and high-manganese steel has a hexagonal lattice; however, this martensitic material appears

to exist only on the surface layer and is probably the result of transformation of the austenite brought about by polishing stresses.

Austenite→ Martensite→ Pearlite. The Classical View. K. Honda. (Metal Progress, 1935, vol. 28, Aug., pp. 34–35). The author discusses the structural changes which occur when the A_1 transformation takes place. He upholds the classical view that martensite is the intermediate product when austenite tranforms into pearlite, and suggests that there is no essential difference between nodular troostite and troostite obtained by tempering martensite.

The Electron-Optical Observation of the Transformation of Iron from the α to the γ Condition. E. Brüche and W. Knecht. (Zeitschrift für technische Physik, 1935, vol. 16, pp. 95–98). By means of the electron-microscope, the authors have determined that very pure electrolytic iron recrystallises at 898° C. on cooling and at 907° C. on heating. The addition of 1 per cent. of silicon raises the critical temperatures by about 20°.

The Mechanism of the α–γ Transformation in Iron. G. Wassermann. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1935, vol. 17, No. 11, pp. 149–155). The author first discusses in detail the phenomena of recrystallisation that occur in unalloyed low-carbon iron when heated above the A_3 point. He then describes the results of an investigation of the α–γ transformation in a nickel-iron alloy containing 30 per cent. of nickel. Comparison of the relationship between the orientation of the γ and α phases in this alloy and in martensite containing carbon led him to postulate a slightly different relationship for the transformation into a cubic body-centred phase, which fitted the experimental observations better. The corresponding mechanism of the transformation is simpler and clearer than that proposed by Kurdjumow and Sachs.

Investigations on the Heat Evolutions at the Transformations in Iron. H. von Steinwehr and A. Schulze. (Zeitschrift für Metallkunde, 1935, vol. 27, June, pp. 129–132). **Heat Evolution at the A_2 and A_3 Transformations in Pure Iron.** H. von Steinwehr and A. Schulze. (Zeitschrift für Physik, 1935, vol. 36, p. 419: Stahl und Eisen, 1935, vol. 55, June 27, pp. 705–706). The authors have determined the heat evolution at the A_2 and A_3 points in very pure electrolytic iron (99·97 per cent. pure), containing only 0·03 per cent. of manganese, traces of silicon and no nickel. The measurements were made in a heat-insulated calorimeter; cooling curves were first taken, from which the rise of temperature at the transformation point was calculated, and from this the heat evolution was deduced as the product of this temperature rise and the total heat capacity of the whole apparatus at the transformation temperature. The mean value of the heat

evolution at the A_2 point was found to be 4·8 cal. per grm., with an average error of ± 0·2 cal. per grm. The first signs of the transformation appeared at above 790° C. and the last were still observable at about 710°. The A_3 transformation commenced at 894° and ended at 870°; contrary to the A_2 point, a distinct arrest was observed at A_3. Consequently the simple formula for calculating the heat evolution could not be used, but recourse was had to the general equation, from which the cooling constants for Newton's law and the area of the transformation range had to be calculated first. The mean value of the heat evolution at A_3 was found to be 6·2 cal. per grm., with an average error of ± 0·08 cal. per grm.

A Dilatometric Study of the Alpha-Gamma Transformation in High Purity Iron. C. Wells, R. A. Ackley and R. F. Mehl. (American Society for Metals, 1935, Preprint No. 5). The authors have made a dilatometric study of the α–γ transformation in high-purity iron, using a newly developed high precision recording differential dilatometer. About sixty heating and cooling dilatation curves were made on hydrogen-purified Armco iron, electrolytic iron, and carbonyl iron. The temperature of the α–γ transformation was fixed at 909·5 ± 1° C. Thermal analysis curves were obtained simultaneously, and observations on the effects of heterogeneity, on anomalous length changes, on degrees of undercooling and superheating, and on the speed of this transformation are reported.

The Influence of the Alloying Elements on the Transformations of the Austenite and the Strength Properties of Alloy Steels on Stepped Heat Treatment. H. Döpfer and H. J. Wiester. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, June, pp. 541–546: Technische Mitteilungen Krupp, 1935, vol. 3, No. 3, pp. 87–99). The course of the transformations in the undercooled austenite in low-alloy chromium, silicon, molybdenum, tungsten, vanadium, manganese and nickel steels were observed magnetometrically. In all these steels the rate of transformation rose above the martensite point to a maximum lying between about 300° and 400° C. In the chromium, molybdenum and vanadium steels it then fell to a range of high stability of the undercooled austenite at about 500° C., and only rose again in the pearlite range, whereas in the nickel, manganese, silicon and tungsten steels the range of the intermediate stage passed directly into the pearlite range. The occurrence of a range of stability in chromium-nickel steels, such as Wever, Lange and Jellinghaus had investigated, was therefore to be attributed to the chromium content. Frequently the transformation of the austenite set in first after a certain period, which could be of considerable length, particularly at low temperatures and in the stable range at 500° C., and which could also be markedly influenced by the content of alloying element

and also of carbon. The initial period and the maximum rate of transformation together give an indication of the conditions to be maintained in practical step-wise hardening and heat treatment. Tests with various alloyed specimens showed that under favourable conditions step-wise heat treatment could produce mechanical properties as good as those obtained by ordinary heat treatment. Step-wise treatment, however, should only be used in practice in special cases, as a means of avoiding stresses and hardening cracks.

Initiatory Period of the Austenite Transformation. E. Scheil. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, June, pp. 565–567). The possibility of suppressing transformations is determined by the time required for a nucleus of the transformation product to form. This initiatory period can be calculated from the relationship between the temperature of the transformation and the rate of cooling. The relation of the initiatory period to the temperature has a minimum, the exact position of which is not yet known in detail. Measurements made by Sato have been used to draw up a layer-line diagram, in which a ferrite, a cementite and a pearlite field occur. The formation of a pearlite double nucleus occurs more easily in the pearlite field than that of a single nucleus of ferrite or cementite. The effect of already present ferrite or cementite crystals is small; it is overshadowed by the influence of the carbon content of the austenite on the formation of nuclei. The sequence of the inocculating actions of various substances on ferrite is as follows: Sulphides; grain boundaries of austenite; oxides.

The Transformation Temperatures of Martensite and Austenite in Hardened Steels. B. Kjerrman. (Jernkontorets Annaler, 1935, vol. 119, No. 6, pp. 217–231). By measuring the change in electrical conductivity the decomposition of tetragonal martensite and of austenite at various constant temperatures and with slowly rising temperature was studied. The steel used contained approximately 1 per cent. each of carbon, manganese and chromium, and 0·5 per cent. of silicon. With a moderate rise in temperature the martensite was found to decompose between 110° and 150° C., and the austenite between 240° and 290° C. By keeping the temperature constant for long periods of time decomposition of the martensite was observed to set in quite early. From the results of an experiment in which a specimen was tempered at 150° C. for 1 hr. and afterwards maintained at 120° C. for 280 days, it is concluded that austenite cannot begin to decompose until the tetragonal martensite has almost completely disappeared, but that when that has taken place the austenite decomposes slowly even at as low a temperature as 120° C.

Initial Stages of the Magnetic and Austenite Transformations

in a Carbon Steel. I. N. Zavarine. (American Institute of Mining and Metallurgical Engineers, 1935, Oct., Technical Publication No. 646). Experiments were performed on high-carbon wires (approximately 0·9 per cent. carbon content) in which the wires were heated to quenching temperature, quenched in a bath of molten salt, passed through an X-ray camera and finally passed through a transformer which served to indicate the magnetic state of the wire. It is shown that the initial stages of magnetic transformation in medium- and high-carbon steels quenched in a hot bath are delayed for a measurable time, depending upon the quenching conditions and the carbon content of the steel. The lattice transformation of a eutectoid steel quenched in a hot bath does not coincide with the magnetic transformation, but precedes it.

The Iron-Cobalt-Tin System. W. Köster and W. Geller. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, June, pp. 557–560). The structure of the ternary iron-cobalt-tin system with up to 40 per cent. of tin is outlined ; transformations between the individual iron-tin compounds and the γ–α transformation of cobalt are disregarded. The equilibrium α solid solution + melt = Fe_2Sn moves to higher temperatures as cobalt is added. It combines with the markedly sinking equilibrium, α solid solution + melt = γ solid solution of the iron-cobalt system and the slightly sinking equilibrium, melt = γ solid solution + Co_2Sn to form a four-phase transition equilibrium, melt + γ solid solution = α solid solution and compound solid solution. The phase fields in the solid condition were determined by measurements of the hardness, coercive force and the loss of magnetism, and also by microscopical observations.

Contribution to the Knowledge of the Transformations in Irreversible Iron-Manganese Alloys. E. Scheil. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Aug., pp. 115–116). By measurements of the change of length of iron-alloy specimens containing from 7 to 12 per cent. of manganese, it was shown that an extension marked on the heating curve of manganese steels, which H. Scott had assumed to represent a γ–α transformation, was really due to a ξ–γ transformation. The two transformations, of which the γ–α is associated with an extension and the γ–ξ with a shortening of the specimen, overlap in a range between 7 and 12 per cent. of manganese ; the transformation temperatures on cooling and on heating were determined.

Alloys of Iron, Manganese and Carbon.—Part XIV. Iron-Carbon Alloys containing Seven Per Cent. Manganese. C. Wells and F. M. Walters, jun. (Transactions of the American Society for Metals, 1935, vol. 23, Sept., pp. 751–760). Pure iron-manganese alloys containing approximately 7 per cent. of

manganese and carbon varying between 0·02 and 1·25 per cent. were studied and a constitutional diagram was drawn. The alloys, brought to substantial equilibrium at various temperatures, were quenched and subsequently examined under the microscope. (*See* Journ. I. and S.I. 1934, No. II. p. 704).

The Irreversibility of the Iron-Nickel Alloys and their Equilibrium Diagram. E. Scheil. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, Sept., pp. 163–166). The range of irreversibility of nickel steels has already a notable width in the region of noticeable atomic movement. In this temperature range a transformation can occur over the entire irreversibility range. It is suggested that the cause of the occurrence of a range of irreversibility is the coupling of the rapidly-proceeding lattice changes with a slow alteration in the composition. Two new points for the so-far quantitatively undetermined equilibrium diagram have been obtained; with their help the supposed temperature of the α–γ equilibrium in the range 0 to 10 per cent. of nickel has been given.

Knowledge on the Iron-Nickel Alloys. O. Dahl, J. Pfaffenberger and N. Schwartz. (Metallwirtschaft, 1935, vol. 14, Aug. 23, pp. 665–670). The limits of the miscibility gap in the copper-nickel-iron tenary system for the temperature range below 1,050° C. are given; at 1,050° it reaches to about 43 per cent. of nickel, while at room temperature and at that of incipient diffusion it extends to about 65 per cent. of nickel Thus the large majority of these alloys fall in the group of hardenable alloys. The mechanical hardenability is not great—the alloys can be deformed and machined easily in both the quenched and the hardened condition; on the other hand, the magnetic hardenability is very great, particularly in a region towards the nickel-copper side. Those alloys which have a high magnetic permeability after rapid cooling can be used for permanent magnets. Although the remanence is small on account of the low iron content the product of the remanence and the coercive force is equal to that for cobalt steels. These alloys are distinguished from all other permanent-magnet alloys by their easy deformability. The changes of properties brought about by heat treatment are described. The condition of precipitation corresponding to maximum magnetic hardness is not attained until long after the condition of maximum change of resistance has been reached.

Contribution on the System Iron-Phosphorus-Oxygen. H. Wentrup. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, July, pp. 57–60). The author has determined the constitutional diagram of the system Fe_2O_3–P_2O_5 between 30 and 50·6 per cent. of iron, and that of the system FeO–P_2O_5 between 46·9 and 59·8 per cent. of iron. It was found, in particular, that the compounds $Fe_2O_3.P_2O_5$, $2Fe_2O_3.3P_2O_5$ and $3FeO.P_2O_5$ were

stable up to the melting point. The conjectural constitutional diagram for the two-component system $3FeO.P_2O_5$–$2Fe_2O_3.3P_2O_5$ was also drafted.

Equilibrium Diagrams of Binary Alloys of Tin. E. S. Hedges and C. E. Homer. (Technical Publications of the International Tin Research and Development Council, 1935, Series B, No. 2). The more important equilibrium diagrams relating to the binary alloys of tin are presented in convenient form, and a guide to the interpretation of the diagrams is included for the use of those who may not have previously studied the subject academically.

Notes on the Solidus Temperatures in the Systems Iron-Tungsten and Iron-Molybdenum. W. P. Sykes. (American Society for Metals, Oct., 1935, Preprint No. 7). By observations on the melting temperatures of wire specimens the solidus lines in the iron-rich portions of the iron-tungsten and iron-molybdenum binary systems were determined. Microstructures of small block specimens drastically quenched indicate that in both systems the α-phase is formed as the product of a peritectic reaction thus: liquid iron $+ \xi \rightleftharpoons \alpha$.

The Crystal Structure and the Composition of the Intermediate Phases present in Iron-Tungsten and Iron-Molybdenum Alloys. H. Arnfelt and A. Westgren. (Jernkontorets Annaler, 1935, vol. 119, No. 5, pp. 185–196). The authors have determined the crystal structure of the intermediate phases present in iron-tungsten and iron-molybdenum alloys, on the basis of X-ray analyses carried out by Arnfelt (Iron and Steel Institute, Carnegie Scholarship Memoirs, 1928, vol. 17, p. 1). The phases present in commercial ferro-tungsten are chiefly Fe_2W and tungsten; Fe_7W_6, although a possible constituent, appears to be unstable at low temperatures; on the other hand, iron-molybdenum alloys contain Fe_7Mo_6 as the sole intermediate phase, and this is stable even on prolonged heating.

An Investigation of the Zinc-Rich Portion of the System Iron-Zinc. E. C. Truesdale, R. L. Wilcox and J. L. Rodda. (American Institute of Mining and Metallurgical Engineers, 1935, Oct., Technical Publication No. 651). Zinc-rich alloys were investigated by the usual methods of thermal analysis, microscopical examination, electrical and X-ray measurements, and so on, in order to re-determine the liquidus and adjacent parts of the diagram. The results, and a description of the technique are contained in Part I. of the paper. In Part II. an attempt is made to review all the published work bearing on the constitution of the iron-zinc system and to correlate previous work done by the authors.

The Zinc-Iron Equilibrium Diagram. W. D. Jones and R. T. Parker. (Metallurgist, 1935, Aug., pp. 56–58). The work of

numerous investigators on the iron-zinc diagram is critically reviewed.

HEAT-RESISTANT AND ACID-RESISTANT ALLOYS.

The Further Development of Steels Resistant to the Atmosphere. K. Daeves. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, July, pp. 37–40). The author presents the results of tests on the natural rusting, in industrial and rural atmospheres, of steels containing up to 0·2 per cent. of phosphorus and up to 0·3 per cent. of tin, in addition to 0·3 per cent. of copper. Contrary to the indications of solution tests, even in a variety of acids, a phosphorus content exceeding 0·06 per cent. increases the resistance of copper-bearing steel to the ravages of the weather, irrespective of the process by which the steel was made; the resistance to rusting of copper-tin steels is likewise improved. On copper steels the higher phosphorus content causes the rapid appearance of a dark-brown, smooth, strongly adherent protective layer of rust, free from pits. As a result of this freedom from pits, the raised phosphorus content is of benefit in maintaining the strength of such materials as fencing wire. Puddled steel without copper was the equal of copper steel only under conditions of comparatively slow rusting—for instance, in rural situations. The blue rolling-mill scale weathered off copper steels with raised phosphorus contents more rapidly than off normal steels; also, the former steels behaved better under paint coatings than did the latter.

Resistance Alloys for Heating Elements. (Metallurgist, 1935, June, pp. 35–37). The work of Schulze and Hoffmann, and of Grunert, Hessenbruch, and Schichtel on the melting points of alloys used for heating elements is reviewed. It is shown that the melting point of aluminium-chromium-iron alloys such as Megapyr and Kanthal is 1,500° C.—a value less than that generally claimed for these materials. It would appear that though aluminium-chromium-iron elements are superior to nickel-chromium alloys at temperatures above 1,200° C., they have certain disadvantages; thus they are not rustless, and are brittle when cold.

Electrical Resistance Alloys: Why Nickel-Chromium so Successfully Serves as Heating Element Material. F. P. Peters. (Electrochemical Society, 1935, Preprint No. 68–27). The six fundamental properties required of a high-temperature metal resistor (high resistivity, low temperature coefficient of resistivity, resistance to oxidation and scaling at temperatures up to 1,150° C., high melting point, strength and physical stability at high temperatures and reasonable cost) are discussed at length. Several resistor materials are compared. It is shown that the 80/20

nickel-chromium alloy is the most suitable; the oxide which forms on the surface has a coefficient of temperature expansion similar to that of the wire itself, while the latter is not subject to severe growth, resists high temperatures and is not too costly.

A New Heat-Resistant Alloy. S. L. Hoyt and M. A. Scheil. (American Society for Metals, Oct., 1935, Preprint No. 19). **A New Resistor Alloy – Chromium-Aluminium-Iron.** S. L. Hoyt. (Metal Progress, 1935, vol. 28, July, pp. 38–40). **Heat-Resisting Alloy.** (Steel, 1935, vol. 97, Sept. 2, pp. 41–42). The authors describe a new heat-resistant iron-chromium-aluminium alloy, known as Smith Alloy No. 10, and containing 37·5 per cent. of chromium, and 7·5 per cent. of aluminium. It is melted in a high-frequency furnace, and is produced in the form of rod, ribbon or wire for use in electric-furnace resistor elements.

Chromium Alloy Steel. (Steel, 1935, vol. 97, Aug. 5, pp. 34–35). A new alloy, of which a hopper car and a merchandise container have been constructed, is described. Known as RR-11, the alloy has been introduced by the Rustless Iron Corp. of America, Baltimore. It contains 11 per cent. of chromium and 0·05 to 0·07 per cent. of carbon, and is said to yield the following physical properties: Ultimate tensile strength, 90,000 lb. per sq. in.; yield point 65,000 lb. per sq. in.; elongation on 2 in. 15 to 25 per cent.; hardness, 185 Brinell: and Izod value, 90 ft.-lb. The alloy is produced in rolled sheets which are corrosion-resistant and easily hot- and cold-worked, welded and machined.

Notes on 29–9 Alloy. R. J. Wilcox. (Metal Progress, 1935, vol. 27, May, pp. 49–51, 66). The 29—9 alloy contains: Carbon 0·20–0·30; manganese 0·55–0·75; silicon 0·75–1·25; nickel 8·00–10·00; chromium 28·00–30·00; sulphur 0·05 (max.); phosphorus 0·05 (max.) per cent. The alloy consists of ferrite, austenite, and free carbides; unlike the 18 8 chromium-nickel alloy, the as-cast material is not susceptible to intergranular corrosion; this is a great advantage in the case of parts which cannot be heat-treated. The material has tensile properties similar to the 18 8 alloy, and it can be readily welded. For high-temperature service the alloy is made with higher nickel and carbon contents; its performance may be compared with the 35 per cent. nickel, 15 per cent. chromium alloy often employed.

Nickel-Chromium Heat Resisting Alloys. A. M. C. Murphy. (Metal Industry, 1935, vol. 47, Oct. 18, pp. 387–390, 402). Of the alloys of iron, chromium and nickel which are used for electrical resistance wire two are of general importance; they are the binary alloy containing 80 per cent. of nickel and 20 per cent. of chromium, and the ternary alloy containing nickel 65, iron 20, and chromium 15 per cent. Both have a high specific resistance —about 110 microhms per centimetre cube. Care must be taken in the production of the 80/20 alloy, and the metal is usually

melted in a high-frequency furnace. Rolling is difficult and is usually preceded by forging; wire-drawing follows ordinary practice but requires completely pickled wire and skill in drawing. The annealing which follows confers a thin oxide coating on the surface of the wire and renders it resistant to high temperatures subsequently. Before being marketed specimens of the wire are subjected to an accelerated life test, during which the wire is constantly heated and allowed to cool.

The Application of Corrosion-Resistant Steels in Chemical, particularly the Nitrogen, Industry. P. Schafmeister and F. K. Naumann. (Technische Mitteilungen Krupp, 1935, vol. 3, No. 3, pp. 100–107). In the nitrogen industry, three broad types of structural steel are used: (*a*) Hydrogen-resistant steels for ammonia synthesis plant; (*b*) rust-, acid- and heat-resistant steels for the manufacture of nitric acid by the oxidation of ammonia; and (*c*) steels resistant to ageing and caustic alkalis for the production of nitrates from nitric acid and bases. The authors discuss the properties of these various steels. Examples of plant made of these materials are illustrated.

Design and Selection of Heat Enduring Alloys for Industrial Furnaces. R. S. Brown. (Industrial Heating, 1935, vol. 2, Sept., pp. 495–498, 514). Attention is chiefly devoted to the proper design of ribbed and similar sections (in view of expansion and other thermal effects), the use of creep stress data, and the properties—such as resistance to oxidation, structural stability and creep strength—which heat-resistant alloys are required to possess.

Chromium Steels of High Nitrogen Content. R. Franks. (American Society for Metals, Oct., 1935, Preprint No. 30). The author discusses the influence of nitrogen in steels. Observations indicate that the marked refinement of grain imparted to high-chromium steels is brought about by nitrogen that is believed to be in combination with chromium as a nitride. Chromium nitride representing only a few hundredths of 1 per cent. greatly increases the hardening capacity of low-carbon, 12 to 14 per cent. chromium steels, a good indication that this nitride promotes austenite formation. The effect of this high solubility of the chromium nitride in low-carbon iron of greater chromium content is far more pronounced, and in this respect the chromium nitrides seem to function in a manner similar to that of the carbides in ordinary steel. The addition of nitrogen to these low-carbon high-chromium irons represents, therefore, a metallurgical procedure designed to limit the grain-size and to improve the strength and toughness without unduly increasing hardness associated with brittleness.

CORROSION OF METALS.

(For Coating of Metals see p. 448.)

The "Probability" of Corrosion. R. B. Mears and U. R. Evans. (Transactions of the Faraday Society, 1935, vol. 31, Mar. pp. 527–542). Measurements were made of the influence of sixteen external variables—such as time, area, oxygen concentration, temperature—on (*a*) the probability, and (*b*) the conditional velocity of corrosion by drops. Oxygen concentration depresses the probability but increases the conditional velocity; this is also true of certain inhibitors. A rise in the temperature of pre-exposure (to dried air) tends to diminish the probability, whereas a rise in the temperature of the experiment greatly increases the probability, as does the presence of sulphur dioxide in the gas phase. The probability is depressed by the presence of neighbouring corroding points. It is shown that, assuming attack to be stifled by the corrosion product formed, the oxygen concentration needed to prevent corrosion in a drop is a suitable measure of the size of the biggest pore.

Effect of Composition on the Corrosion Probability of Iron and Steel. R. B. Mears. (Iron and Steel Institute, Carnegie Scholarship Memoirs, 1935, vol. 24, pp. 69–85). The author points out, in his introduction, that both the probability of corrosion and the conditional velocity are dependent on external variables (such as the surrounding media) and on internal variables (including the concentration and distribution of minor constituents, and the physical condition of the material). The general technique employed in estimating the corrosion probability is briefly considered.

The results obtained from the examination of scale-covered materials are discussed, and it is shown that continuous scales give rise to low probability figures, whilst imperfect scales give higher values. Impurities in the steel are found only to confer special susceptibility on *smooth* metal surfaces, while the presence of particles of certain materials, such as ferrous or manganese sulphide, in aqueous contact with the surface, increases the probability of corrosion. Certain other materials are inert.

The correlation between the corrosion probability and the amount of impurity was investigated for the impurities usually present in ferrous materials. A fairly high degree of positive correlation was indicated for sulphur, carbon and manganese, but there was only a small degree of correlation between the silicon, phosphorus or copper content and the corrosion probability.

The Electrochemical Potential of Iron and Corrosion Phenomena. N. Nekrassow, J. Stern, and Z. Gulankaja. (Zeitschrift

für Elektrochemie, 1935, vol. 41, pp. 2–9). The change of the potential of iron with time in solutions of various salts was compared with the progress of the corrosion. Shifting of the potential of the iron towards the hydrogen side was always connected with an increase in the rate of corrosion, while a shift towards the oxygen side was related to a retardation in the corrosion rate. The hydrogen-ion value and the anion present have an essential influence on the phenomena, and this is discussed in detail.

Fluctuations of the Potential of Iron in Nitric Acid. M. Karschulin. (Zeitschrift für Elektrochemie, 1935, vol. 41, pp. 224–229). The periodic solution of Armco iron in nitric acid solutions of different strengths was observed by means of potential/time measurements. The changes of potential were related to the surface condition of the metal. At the lower potential the iron was covered with a reddish-brown layer. The potential then rose with time to positive values, till suddenly gas evolution commenced, when the highest positive value was reached. The cycle repeated itself at regular time intervals.

The Theory of Passivity Phenomena. XXVI.—The Limiting Currents during the Anodic Polarisation of Metals in Aqueous Solutions. W. J. Müller. (Zeitschrift für Elektrochemie, 1935, vol. 41, Aug. 31, pp. 83–86). The theories of E. Müller and of the present author regarding passivification during anodic polarisation are contradictory. An experimental test of the two points of view is possible in that, according to the author's coating theory, the onset of saturation of the salt solution during the anodic electrolysis of a metal has no noticeable influence on the condition of polarisation of the electrode, and an increase in the polarisation is essentially connected with a precipitation of the solid salt, whereas, according to E. Müller, additional polarisation sets in at the onset of saturation, which causes a diminution of the current strength. Current/time curves, taken during the polarisation of iron in N H_2SO_4 and in the same acid saturated with ferrous sulphate, were found to be continuous, and hence they do not support E. Müller's opinion.

The Nature of Corrosion Phenomena with Oxygen Depolarisation. G. W. Akimow, S. A. Wruzewitsch, and H. B. Clark. (Korrosion und Metallschutz, 1935, vol. 11, July, pp. 145–156). The authors have worked out a theory of the electrochemical cell and have established basic formulæ giving the relationship of the potential difference of the working cell at a given moment, the amount of hydrogen collected on the cathode at that moment, and the current, to the time. Theoretical consideration of the phenomena in the electrochemical cell with oxygen depolarisation, and analysis of the solved equations gave information regarding the conditions on reaching the steady state. That

the current was independent of the resistance, within wide limits, and of the initial potential difference was proved experimentally; the limit of this independence of the resistance was found to be lower than the calculated value, which was attributed to the difficulty of removing the last of the hydrogen from the cathode. Depolarisation was more intense on a surface which had been already attacked than on a fresh one. The influence of the surface areas of the anode and cathode on the current (and the corrosion) was examined anew; the observed relations between the current and the anode and cathode potentials are explained on the basis of the authors' theory, which gives an analytical expression for the curve relating the cathode potential to the extent of the anode and cathode areas. The theory is applied to the explanation of various corrosion phenomena with oxygen depolarisation, such as structural corrosion, resulting from cathode inclusions with low initial potentials, intercrystalline corrosion, enhanced destruction of riveted and welded joints, the action of protectors on distant structural parts, &c.

Corrosion Testing Methods. H. E. Searle and F. L. LaQue. (American Society for Testing Materials, 1935, Preprint No. 29). The authors draw attention to the practical advantages of field corrosion tests and describe a simple and inexpensive device known as the spool-type specimen holder. The holder is applicable to a broad range of field testing, including the exposure of different metals, either plain or welded, and either out of contact with one another or in galvanic combination, to wet solids, liquids or vapours.

Continuous-Flow Corrosion Tests on Steel Pipe. H. S. Rawdon and L. J. Waldron. (American Society for Testing Materials, 1935, Preprint No. 28). The shortcomings of ordinary laboratory corrosion tests, particularly for predicting what may be expected to occur in service, are well recognised. The limitations imposed by the small size of the usual specimen, the relatively short duration of the test and the failure to simulate the essentials of the conditions expected in service, are often very serious ones. Exposure tests, on the other hand, are not always practicable, and the possibility of developing a testing procedure substantially free from the really serious shortcomings of ordinary laboratory corrosion tests remains still a subject of live interest. The method described in the paper has been used with ferrous pipe material and consists essentially in circulating water of known characteristics through a system built up of sections of the pipe to be tested. A number of specimens, approximately 5 in. long, after being cleaned and weighed, were assembled, with intermediate hard rubber spacing rings, to form a vertical water-tight column. After being subjected to water flowing at a controlled rate for the desired period, the columns were dissembled, the specimens

properly cleaned and weighed, and the corrosion rate was determined. Results obtained on 2-in. pipe of two open-hearth steels showed that: (*a*) The distribution of corrosion varies decidedly with the degree of turbulence in the flow; (*b*) the corrosion of specimens varies with their position in the column, the rate being a maximum at the inlet end with a progressive decrease along the column, which effect can be largely overcome by increasing the rate of flow; (*c*) the pipe used successively from one series of tests to another, the specimens being cleaned at the end of each run, showed an apparent increase in corrosion rate with increased use, attributable to roughening of the surface—a factor which is not usually taken into consideration; (*d*) the distribution of corrosion is intimately related to the scale deposition accompanying corrosion.

The Influence of Certain Organic Hydroxy-Compounds on the Corrosion of Iron in Oxygenated Salt Solutions. W. S. Patterson and R. C. A. Culbert. (Journal of the Society of Chemical Industry, 1935, vol. 54, Oct. 4, pp. 327–331T). The corrosion of iron in oxygenated potassium sulphate solution at high concentrations of oxygen produces ferric hydroxide and a thin but visible film of ferrosoferric oxide which adheres to the metal, causing differential aeration. At low concentrations of oxygen, the deposit adhering to the metal is subsequently reduced to ferrosoferric hydroxide. At a low oxygen concentration the rate of corrosion is directly proportional to the concentration of oxygen whilst at higher concentrations the corrosion is more rapid, ultimately becoming constant and independent of further increases in the oxygen concentration. Hydroxy-compounds of the type studied cause retardation of the corrosion, owing, to some considerable extent, to the fact that they are absorbed on the corrosion products and modify both the character of the loose deposit and the black film immediately covering the metal. At higher concentrations they initiate peptisation of the corrosion products and still further reduce the corrosion. The amount of retardation is controlled by three factors: The quantity of inhibitor present, the number of hydroxyl groups per mol of inhibitor and the spatial arrangement of the hydroxyl groups in the molecule.

Corrosion of Steel. T. J. Finnigan, R. C. Corey, and D. D. Jacobus. (Industrial and Engineering Chemistry, 1935, vol. 27, July, pp. 774–780). The corrosion of steel in solutions containing oxygen and carbon dioxide was observed. Definite concentrations of these gases were maintained over a period of 21 days in an apparatus in which water continually flowed over the specimens. During the progress of an experiment the water became saturated with ferric hydroxide. It was found that oxygen and carbon dioxide mutually influence the corrosion rate, 20 parts of carbon dioxide having approximately the same influence as one part of

oxygen. At zero concentration of either gas the corrosion rate is not zero but a value dependent upon the concentration of the other gas. The relation between the corrosion rate and gas concentrations was established in a series of curves.

Corrosion of Metal Combinations and its Prevention. E. K. O. Schmidt. (Metallwirtschaft, 1935, vol. 14, May 24, pp. 409–412). After explaining, briefly, the effect of contact of two dissimilar metals on their respective behaviours in a corroding medium, the author touches on the use of the phenomenon for protecting one metal at the expense of another (zinc coating on steel, zinc plates suspended in boilers, &c.), and discusses the three principal means of preventing the accelerated corrosion of the less noble metal, namely, by avoiding (*a*) the use of different metals, (*b*) the presence of electrolytes, or (*c*) electrical contact or connection between the two metals.

Rusting and Scaling of Structural Steels under Tensile Stress. E. W. Müller and H. Buchholtz. (Archiv für das Eisenhüttenwesen, 1935, vol. 9, July, pp. 41–45). Specimens of commercially pure iron, basic Bessemer and open-hearth steels with 0·07 to 0·16 per cent. of carbon, and a steel with carbon 0·2, copper 0·6 and chromium 0·4 per cent., were allowed to rust for from three to nine months in the atmosphere or in flowing tap water while subjected to tensile stress in varying degree ; the influence on the weight loss and on the tensile properties was observed. Atmospheric attack was not accelerated by tensile stress, even up to 5 per cent. permanent elongation ; in running water, on the other hand, the rusting appeared to increase as a result of stressing. Steels containing (*a*) carbon 0·1 per cent., (*b*) carbon 0·13, molybdenum 0·4 per cent., and (*c*) carbon 0·17, chromium 1, and molybdenum 0·5 per cent., were allowed to scale in air and steam at 600° to 900° C., while under tensile stress. Scaling was accelerated by tensile stresses with permanent elongation only when the layer of scale was brittle and cracked (for instance, at about 600° C.).

Comparative Corrosion Tests on Soft Steel Sheets. F. Eisenkolb. (Korrosion und Metallschutz, 1935, vol. 11, July, pp. 156–162). The author has carried out laboratory corrosion tests on thin sheet of nine steels in order to compare the results with those obtained in exposure tests (*see* Journ. I. and S.I., 1934, No. II., p. 716). After describing the specimens and their preparation, he records their behaviour when submitted to the alternate influence of acid vapours, water and air, and when attacked by various corrosive liquids ; the results are presented in tables and are discussed in detail. It appears that complete parallelism between the laboratory tests and the exposure tests was not obtained, though in certain cases there was some approximation. The steels used could be grouped according to their behaviour into copper-bearing steels, which were very pure, and the usual com-

mercial steels, which contained the normal impurities. In carrying out the tests by the various methods the order in which the corrosive media were applied and their strength were of determinative importance. The principal advantage of copper steels appeared to lie in their use in the open air ; in cold water they lost all their superiority, though they behaved slightly better in warm water. The author concludes by questioning the value of rapid corrosion tests and emphasises that the results obtained must be applied with the utmost circumspection.

Corrosion Tests in Various Refinery Services. J. E. Pollock, E. Camp, and W. R. Hicks. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 639). Results of corrosion tests on a large number of ferrous and non-ferrous materials, under a variety of corrosive conditions—including those met with in refinery practice—are recorded. The relative merits of the alloys used are critically examined, on the basis of the test results obtained.

Effect of Arsenic on the Corrosion of Iron by Sulphuric Acid. O. P. Watts. (Electrochemical Society, Mar. 1935, Preprint No. 17). The author reports the results of tests on the effect of arsenic on the corrosion rate of iron alone and when coupled with other metals in dilute sulphuric acid and in sea water. It is only when it is dissolved in the electrolyte that arsenic inhibits the corrosion of iron, and it has this effect only in solutions in which iron dissolves mainly by visible displacement of hydrogen. It is without effect on the corrosion of iron by oxygen depolarisation.

Effect of Various Elements on the Corrosion Resistance of Mild Steels. T. Murakami, H. Endô, and H. Sekiguchi. (Kinzoku No Kenkyu, 1935, vol. 12, Sept. 20, pp. 430–448). A series of corrosion tests on specimens of low-carbon steels containing small amounts of additional elements, exposed to non-oxidising acid and saline media was carried out. The effect of copper was first studied, the highest corrosion resistance being obtained with the specimen containing 0·35 per cent. of copper, corresponding to the solubility limit in α-iron. The influence of small percentages of various elements on the corrosion resistance of steel with 0·35 per cent. of copper was next examined ; of the elements employed, aluminium, chromium and silicon were found to have a favourable effect, the influence of tungsten, nickel, molybdenum, cobalt, titanium and manganese being either indefinite or unfavourable. Oxidation tests and further corrosion tests were performed on steels containing copper, chromium, silicon or aluminium. These steels exhibit not only high resistance to corrosion in aqueous solutions, but also good resistance to oxidation in air at high temperatures. The authors suggest that this is due to the formation of a stable oxide film on the surface of the steels ; they

further remark that this may inhibit red-shortness in copper steels, since surface oxidation, with its accompanying deposition of the copper-rich phase in the boundaries of the surface crystals, is not likely to take place.

Corrosion of Mild Steels Containing Copper. (Metallurgist 1935, Oct. 25, pp. 79–80). An account of work carried out by various investigators on the corrosion of copper-bearing steels.

Study of the Influence of Cold-Work on the Intergranular Corrosion of 18/8 Rustless Steels. Van den Bosch and Vialle. (Métaux, 1935, vol. 10, Mar., pp. 66–75). The authors first touch on the intergranular changes which take place on reheating annealed 18/8 chromium-nickel steel. They then discuss the changes on reheating the cold-worked material, the mode of precipitation of the carbides, the rate of heating giving rise to the changes, and the influence of cold-work on the temperature range of susceptibility to alteration. Next, they deal with the stabilisation of the steel, starting with cold-worked metal, and its application to welding. Regarding the use of stabilised 18/8 steel they draw the following conclusions: (1) This material can be arc-welded without alteration; (2) when used hot, in contact with reagents capable of setting up intergranular corrosion at temperatures below the temperature of stabilisation, if this stabilisation is not complete the first hours at temperature may cause a precipitation of carbide in the grain boundaries; this precipitation, however, is much less than in annealed 18/8 steel, and consequently homogenisation is more rapid. It is sufficient that the intergranular corrosion due to the reagent shall proceed more slowly than the re-homogenisation.

Marine Corrosion. J. W. Donaldson. (Metallurgia, 1935, vol. 12, Aug., pp. 117–118). In this article the results of investigations by various organisations on the subject of marine corrosion are reviewed.

Resistance of Welds to Corrosion. A. Leroy and M. Bonnot. (Revue de la Soudure Autogène, 1935, vol. 27, Mar., pp. 2–6). The authors first discuss the factors entering into the corrosion of welds and the influence on them of the conditions in which the welds are made. Next they touch on the carrying out of corrosion tests on welds, and then they discuss the corrosion of welds in steel, rustless and semi-rustless steels and non-ferrous metals, on the basis of published papers.

Causes of Corrosion in Air Preheaters. W. Gumz. (Wärmewirtschaft, 1935, No. 6: Fuel Economist, 1935, vol. 10, Aug., pp. 975–976). The author points out that air preheaters of the recuperative type often suffer serious corrosion owing to the deposition of moisture on the metallic surfaces. He then discusses the behaviour of both recuperators and regenerators in

regard to the temperatures which obtain and the likelihood of condensation.

The Control of Corrosion in Air-Conditioning Equipment by Chemical Methods. C. M. Sterne. (American Society for Testing Materials, 1935, Preprint No. 30). Corrosion attack of the metallic parts of air-conditioning systems can be so severe that perforation of ferrous sheets and pipes may take place in as short a time as four months if the wash water or condenser water is not chemically treated. Practically every type of corrosion attack is common in these systems. Service corrosion tests and analyses of the recirculated water show very clearly the causes and degree of severity of that attack. The results of tests indicate the degree of protection that may be obtained by proper chemical treatment. Representative cases are reported.

The Relation between Soil Resistance and Underground Currents. K. G. Lewis and U. R. Evans. (Korrosion und Metallschutz, 1935, vol. 11, June, pp. 121–125). After discussing briefly stray currents from, say, tramway lines, and the paths that they may take, the authors describe laboratory experiments designed to determine the relation between the resistance of the soil and the intensity of the stray currents. Two steel rods were embedded in sand moistened with solutions of various strengths of salt and also tap water ; by suitable electrical circuits it was possible to measure the amounts of current flowing for different values of the resistance when various external e.m.f's were applied. The results obtained are recorded in graphs. When the two rods were embedded at the same depth in the sand or were otherwise arranged " symmetrically," the curves (resistance plotted against the current flowing in the positive or negative direction) were symmetrical about the resistance axis, but when one rod was more deeply buried than the other, or was only half-buried, this was no longer the case, and the current was always stronger when the upper rod was the cathode ; this was due to the fact that the upper rod was better aerated than the lower one. It is concluded that for high resistances, the magnitude of the resistance alone determines the strength of the stray current, but for smaller resistances the degree of access of oxygen has a considerable influence.

Clearance Cavitation in High-Speed Turbo-Machines. H. Mueller. (Zeitschrift des Vereines deutscher Ingenieure, 1935, vol. 79, Sept. 28, pp. 1165–1169). It was found that erosion on the edges of the blades of water turbines without blade rims was caused by cavitation in the clearance space and whirling action against the wall. The flow in the clearance space was therefore investigated, theoretically and practically, and as a result a simple means of preventing erosion due to cavitation in the vortex at the wall is suggested.

LABORATORY EQUIPMENT.

The Building and Public Works Laboratories in Paris. R. L'Hermite. (Génie Civil, 1935, vol. 107, July 13, pp. 29–35). **High-Power Machine for Testing Structural Parts.** R. L'Hermite. (Génie Civil, 1935, vol. 107, July 13, pp. 44–45). The testing laboratories of the Société Mutuelle d'Assurance des Chambres Syndicales du Bâtiment et des Travaux Publiques, situated in Paris, and the equipment installed for carrying out mechanical tests on structural material and parts are described in the first article. In the second, features of the construction of a high-power machine for applying test loads to structural parts are set out ; it comprises five jacks, one of 2,000 tons and the others capable of exerting a force of 100 tons each.

CHEMICAL ANALYSIS.

ANALYSIS OF IRON AND STEEL.

The Determination of Chromium in Stainless Steel, Using Perchloric, Phosphoric and Sulphuric Acids. G. F. Smith and G. P. Smith. (Journal of the Society of Chemical Industry, 1935, vol. 54, June 21, pp. 185T–189T). It is shown that a mixture of 1 volume of 70–72 per cent. perchloric acid and 2 volumes of 85 per cent. phosphoric acid is an excellent solvent for chromium, nickel-chromium and tungsten steels. Solution is complete at 200° C. If a solution of mixed perchloric and sulphuric acids is added to this solvent a method of determining chromium in steels is made possible which is at once economical and rapid. The method gives accurate and reliable results.

Estimation of Iron in the Presence of Titanium : An Aeration Process. W. M. Thornton, jun., and R. Roseman. (Journal of the American Chemical Society, 1935, vol. 57, pp. 619–621).

A Modified Persulphate-Arsenite Method for Manganese. E. B. Sandell, I. M. Kolthoff and J. J. Lingane. (Industrial and Engineering Chemistry, 1935, Analytical Edition, vol. 7, July 15, pp. 256–258). A method for the determination of manganese in steel is proposed, in which the permanganic acid formed by the oxidation of the sample with ammonium persulphate in the presence of phosphoric acid with silver as the catalyst, is titrated with a reducing solution containing equivalent amounts of sodium arsenite and sodium nitrite, instead of with the customary sodium arsenite alone. The heptavalent manganese is reduced to the divalent condition and the solution therefore becomes colourless at the end-point. Small amounts of chromium, vanadium, nickel and molybdenum do not interfere. Silver must be precipitated as the chloride before the titration can be made.

Separation of Molybdenum in Alloy Steels, especially from Tungsten, and its Potentiometric Determination. W. Werz. (Zeitschrift für analytische Chemie, 1935, vol. 100, pp. 241–257).

Colorimetric Methods for the Determination of Phosphorus. C. Zinzadze. (Industrial and Engineering Chemistry, 1935, Analytical Edition, vol. 7, July 15, pp. 227–230). Two colorimetric methods for the determination of phosphorus in the presence of silica, arsenic, iron and nitrates are outlined. The methods depend on the use of molybdenum blue in sulphuric acid, or the reduction of MoO_3 with stannous chloride. The author describes the technique to be adopted in each case.

Sulphur Determination by the Combustion Method in Carbon Steels, High-Speed Steels, 18 : 8, &c. H. A. Kar. (Metals and Alloys, 1935, vol. 6, Aug., pp. 223–225). The sulphur is determined by burning a sample of the steel in a stream of oxygen in a bent-over combustion tube and absorbing the sulphur dioxide and trioxide so produced in sodium peroxide solution. The solution is boiled, acidified with HCl, and the sulphur is precipitated with barium chloride in the usual way.

A Rapid Method for the Determination of Sulphur in Ferromagnetic Alloys. B. L. Clarke, L. A. Wooten and C. H. Pottenger. (Industrial and Engineering Chemistry, 1935, Analytical Edition, vol. 7, July 15, pp. 242–244). Existing methods for the determination of sulphur in iron and steel failed to give satisfactory results when applied to nickel alloys of low sulphur content. A new method was devised which consisted of heating the turnings in hydrogen at 1,100° to 1,200° C., the hydrogen sulphide evolved being collected in ammoniacal cadmium nitrate solution and the resulting cadmium sulphide estimated by the Volhard method, or by a modification of the iodometric method. The precision of the method, for the range 0·005 to 0·020 per cent. of sulphur, is $\pm 0{\cdot}001$ per cent. of sulphur on the basis of a 10-grm. sample.

Electrolytic Determination of Titanium, Zirconium or Aluminium in Chromium Steels and Alloys. H. A. Kar. (Metals and Alloys, 1935, vol. 6, June, pp. 156–157). The sample of chromium-nickel steel is dissolved in 10 per cent. sulphuric acid and filtered into a beaker for the subsequent electrolytic separation. The beaker contains mercury at the bottom forming the cathode ; the solution is constantly agitated during electrolysis by means of a motor-driven stirring rod. The anode is a small platinum plate. The solution is electrolysed at 4 amp. for 1½ to 2 hr. The residue previously obtained is washed, fused with potassium bisulphite, dissolved, and combined with the main solution whilst electrolysis is in progress. This treatment removes interfering elements and leaves the titanium, zirconium and aluminium, if present, in the electrolyte. The solution is then filtered, and the filtrate is either (*a*) treated with cupferron, filtered and the silica volatilised off with HF, the final residue being ignited and weighed, or (*b*) treated with ammonia and ammonium persulphate and filtered, the silica being volatilised off as before, and the residue fused with potassium bisulphate, dissolved and subjected to precipitation by H_2S. This method with suitable modifications can be used for estimating aluminium in nitralloy and similar steels.

Progress in Spectographic Analysis. T. A. Wright. (Metals and Alloys, 1935, vol. 6, Sept., pp. 229–234 ; Oct. pp. 289–292). These articles constitute a correlated abstract of six papers on aspects of spectrographic analysis. They contain notes on the

spectographic analysis of non-ferrous and ferrous materials, sampling, electrodes, and the precision and range of the methods adopted.

Industrial Spectrum Analysis.—I. C. S. Hitchen and C. J. D. Gair. (Chemical Age, 1935, vol. 33, July 20, pp. 51–52). It is pointed out that spectrography is an invaluable method of determining the minor constituents of a sample of material; it is also shown that the phenomenon associated with spectral radiation —namely, the disappearance of the faint lines first as the spectrum becomes weaker—enables very small concentrations to be estimated. Attention is then devoted to the apparatus used, and the authors describe the type of electric arc which is used for the examination of material in the rod, powder or liquid form.

Industrial Spectrum Analysis.—II. C. S. Hitchen and C. J. D. Gair. (Chemical Age, 1935, vol. 33, July 27, pp. 75–76). In the second article on spectrum analysis the apparatus required for spark spectra is described; this consists of a light source (a high-voltage spark discharge) and a light condensing system. Spectographs and spectrometers are briefly touched on.

Industrial Spectrum Analysis.—III. C. S. Hitchen and C. J. D. Gair. (Chemical Age, 1935, vol. 33, Aug. 3, pp. 99–101). In the third article the authors proceed to examine the methods of spectrum analysis commonly employed. They show that for reasonably accurate quantitative work the internal standard method, in which the intensities of the lines due to a minor constituent are matched with lines due to the main constituent, should be adopted, in order to eliminate the effect of variation in the electrical discharge conditions. Applications of the internal standard method, in which homologous pairs in one case, and a logarithmic wedge sector in another, are employed, receive consideration.

Errors in the Quantitative Spectrum Analysis of the Iron-Silicon System. G. Scheibe and A. Schöntag. (Archiv für das Eisenhüttenwesen, 1935, vol. 8, June, pp. 533–540). The authors discuss the possible sources of error in the quantitative spectrum analysis of alloys in the iron-silicon system, and describe their investigation of the direction and extent of their influence. They deal with the "three-line" method and its improvement by the introduction of a platinum "step" immediately in front of the spectroscope slit which reduces the intensity of the lines to about one-quarter over about half their length; in this way only one line of the basis metal and one line of the added element are required for measurement, and, further, the lines can be measured in the weakened or unweakened part, so that the range of utility of a line in regard to the concentration is increased tenfold. The authors show that under the conditions described the average error for steel is $\pm$ 3 per cent. (it may rarely rise to

8 per cent. under very unsuitable conditions), for chill-cast iron the maximum error is ± 9 per cent. and for grey iron ± 11 per cent., for a silicon content of 0·2 to 1·0 per cent. in the iron.

A New Method of Extraction of Gases in Metals. L. Moreau, G. Chaudron and A. Portevin. (Comptes Rendus, 1935, vol. 201, July 16, pp. 212–214; Metallurgist, 1935, Oct. 25, p. 71). A method of extracting gases from metals, in which the specimen is made the cathode of an evacuated discharge tube, is described. The process is carried out in the cold with a current of 5 milliamp. at 150,000 v. So far, tests have been made on aluminium, calcium, iron and magnesium.

ANALYSIS OF FUEL.

British Standard Specification for the Sampling of Coal Tar and Its Products. (British Standards Institution, No. 616–1935).

Variables in Coal Sampling. J. B. Morrow and C. P. Proctor. (American Institute of Mining and Metallurgical Engineers, 1935, Technical Publication No. 645).

Sampling and Samplers. P. Rzezacz. (Glückauf, 1935, vol. 71, July 27, pp. 701–709). The author discusses the errors which may occur in the sampling of coal by various means, and appliances for carrying out the sampling mechanically. In conclusion, he deals with the organisation of the sampling in a washery.

Note on the Determination of Nitrogen in Coal. P. E. Hall. (Journal of the Chemical, Metallurgical and Mining Society of South Africa, 1935, vol. 36, Aug., pp. 28–30). A modification of the Kjeldahl-Gunning-Wilfarth method of estimating nitrogen in coal, due to Beet, is described. In the original method a sample of the coal is digested with sulphuric acid in the presence of a K_2SO_4–Hg catalyst. This is followed by distillation of the ammonia into acid, and back-titration. Beet's modifications include the substitution of a small quantity of selenium for some of the mercury in the catalyst, the absorption of the ammonia by boric acid, followed by an acid titration, and the use of a mixed methyl-red/methylene-blue indicator in this titration. This improved method is accurate, quick, and less costly than the original Kjeldahl estimation.

New Gas Analysis Apparatus. F. Büchler. (Glückauf, 1935, vol. 71, July 6, pp. 641–643). The author gives details of a measuring burette capable of being read to 0·01 per cent. by volume, and a new form of gas absorption vessel.

Accurate Determination of Naphthalene according to Zwieg and Kossendey. H. Winter. (Glückauf, 1935, vol. 71, Aug. 17,

pp. 789–790). A short note on an improved method for determining naphthalene in gases, devised by Zwieg and Kossendey (Gas- und Wasserfach, 1935, vol. 78, p. 101). The difficultly soluble molecular compound formed by naphthalene with picric acid has long been used for this purpose, but to obtain accurate results the operating conditions must be held within very narrow limits, as the temperature and concentration of the picric acid have a very marked influence. In the improved method described the gas passes through a wash-bottle containing $N/10$ sulphuric acid (to remove ammonia), then through the absorption train proper, and then to the volume-meter. The temperature of the absorption train is maintained at 0° C. by placing it in Dewar flasks and surrounding it with broken ice. The naphthalene picrate precipitated in the absorption train is filtered off, washed, and redissolved in hot water; the picric acid thus liberated is titrated with $N/10$ alkali solution, using phenolphthalein as indicator.

NOTICES OF RECENT BOOKS PRESENTED TO THE LIBRARY.

CLEAVES, H. E. and J. G. THOMPSON. "*The Metal—Iron.*" (Alloys of Iron Research, Monograph Series, Frank T. Sisco, Editor). 8vo, pp. 574. Illustrated. New York and London, 1935: McGraw-Hill Book Company, Inc. (Price 36s.)

This text book forms the sixth volume of the Monograph Series of the Alloys of Iron Research published by the Engineering Foundation. It reviews all the available information on the preparation and properties of metallic iron of high purity, obtained by the two principal methods, namely, by electrolytic deposition, and by chemical methods. It has never been possible to determine exactly the fundamental properties of absolutely pure iron, that is, of iron which contains no trace of any other element. Such iron has never been prepared, at least in useful quantities, and the best available approximation of the properties of pure iron can only be made by a study of the properties of the purest irons on record. The authors accordingly have defined these as the properties of high-purity iron, and throughout the book the expression, pure iron, is avoided. The succeeding chapters deal with the structure, thermal properties, electric and magnetic properties, the chemical and mechanical properties, the question of heat treatment and ageing, and the effect of minor constituents. An appendix to the volume contains a condensed summary of all the physical, chemical and mechanical properties of iron of the highest attainable purity, some of the values probably representing those of absolutely pure iron. At the end is given a practically complete bibliography of the literature on the subject, arranged in chronological order, dating from 1827 to 1935, and followed by an index of authors' names in alphabetical order. The whole work forms a most useful survey of the knowledge and characteristics of the metal iron, so far as these have been able to be ascertained with the means at present at disposal.

DWYER, P. "*Gates and Risers for Castings.*" Function and Application of Gates and Risers in the production of castings of grey iron, brass, bronze, aluminium, steel and malleable iron. First edition. 8vo, pp. 364. Illustrated. Cleveland, Ohio, and London, 1935: The Penton Publishing Company. (Price 13s.)

This book has been written primarily for the benefit of foundrymen who may not have full opportunity to study the art of foundry practice in all its numerous ramifications. There is no standardised practice among foundrymen in the proper application of gates and risers, and no type of gate has been invented to serve satisfactorily and universally on all classes of castings; under different conditions a particular gate may be favoured by one man and absolutely condemned by another. In this work specific instances are presented of satisfactory methods of gating for practically all classes of castings, and a fairly successful effort has been made to explain the principles of their application and how such principles are affected by the many variable factors incident to foundry operation. No attempt has been made to investigate abstruse theories on physical, mechanical, or metallurgical phenomena, the author's intention having been to compile a practical treatise for the use and benefit of men confronted with the

solution of the every-day problems arising in the course of their work. Throughout a period of 40 years the author has had unlimited opportunity to study foundry methods on every type of casting, and the present volume is based on his accumulated experience acquired during the pursuit of his activities as a practical foundryman.

EGGERT, J., und E. SCHIEBOLD. "*Anwendungen der Röntgen- und Elektronenstrahlen mit besonderer Berücksichtigung organisch-chemischer Probleme.*" (Röntgentagung in Bonn, 1934). Herausgegeben im Auftrage der Deutschen Gesellschaft für technische Röntgenkunde beim Deutschen Verband für die Materialprüfungen der Technik. (Ergebnisse der technischen Röntgenkunde. Band IV.) 8vo, pp. vii + 190. Illustrated. Leipzig, 1934: Akademische Verlagsgesellschaft, m.b.H. (Price 19.50 marks.)

This publication consists of a collection of monographs chiefly dealing with the practical application of X-rays and electron rays to the study of problems concerned with organic chemistry. The papers in question were presented at the annual meeting of the Deutsche Bunsengesellschaft held in Bonn on May 19, 1934, and they have now been issued to form Volume IV. of the valuable series of Proceedings of the Deutsche Gesellschaft für technische Röntgenkunde. They relate largely to the investigation of organic structures, such as cellulose and its derivatives, graphite, &c. Modern X-ray tubes are described by A. Bouwers, and W. Gerlach gives a short history of Röntgen's discovery of the X-rays.

EGGERT, J., und E. SCHIEBOLD. "*Anwendungen der Durchstrahlungsverfahren in der Technik.*" Herausgegeben im Auftrage der deutschen Gesellshaft für technische Röntgenkunde beim Deutschen Verband für die Materialprüfungen der Technik. (Ergebnisse der technischen Röntgenkunde. Band V.) 8vo, pp. vii + 118. Illustrated. Leipzig, 1935: Akademische Verlagsgesellschaft, m.b.H. (Price 15 marks.)

The present publication forms volume V. of the proceedings of the "Deutsche Gesellschaft für technische Röntgenkunde beim Deutschen Verband für die Materialprüfungen der Technik" (Committee 60 of the D.V.M.). It consists of a series of 10 papers presented to the German Society for X-Ray Technology on the occasion of its Spring Meeting at Leipzig, 1935, the authors and titles of which are enumerated below:

R. Berthold. Der Augenblickliche Stand der Röntgen- und Gammadurchstrahlung. (The present position of Röntgen- and Gamma radiography.)

M. Widemann. Die Sonderstellung der Durchstrahlungs verfahren im Werkstoff-Prüfwesen. (The special Rôle of Radiographic Methods in the Testing of Materials.)

P. Brenner. Röntgenprüfung in der Luftfahrt. (X-Ray Testing as applied to Aircraft.)

E. Schatzmann. Röntgenprüfung in der Marine. (X-Ray Testing in the Navy.)

A. Matting. Röntgenprüfung im Werkstoff-Prüfwesen der Reichsbahn. (X-Ray Testing, as applied in Testing Methods for Railway Material.)

W. Schmidt. Röntgenprüfung beim Brückenbau. (X-Ray Testing in Bridge Construction.)

F. Stäblein. Durchleuchtung eines Druckbehälters für die chemische Industrie mit Mesothorstrahlung. (Irradiation of a Pressure Tank for the Chemical Industries by means of Mesothorium Radiation.)

O. Vaupel. Röntgenprüfung von Stahlseilen. (X-Ray Testing of Steel Ropes.)

W. Grimm. Technische Hilfsmittel für die praktische Ausführung von röntgentechnischen Grobstrukturuntersuchungen. (Technical Aids for the Practical Performance of X-Ray Investigations of Coarse Structures.)

H. Stintzing. Stand der Technik offener Hochleistungs-Röhren für die Röntgen-Durchleuchtung. (The Position with regard to the Technology of Unsealed High-Duty Tubes for X-Ray Penetration.)

"*International Repertory of Centres of Chemical Information.*" La. 8vo, pp. 115. Paris: Office International de Chimie.

The function of the International Office of Chemistry is not itself to undertake any work of documentation but to direct the student to the sources likely to afford him the desired information. The Repertory of International Centres of Documentation is the outcome of a resolution passed by the International Union of Chemistry at its Fourth Conference at Cambridge, to the effect that an enquiry should be set on foot with regard to existing centres of documentation, with a view to the publication of a list of various offices in all the principal countries which issue abstracts and bibliographies of literature dealing with chemistry and allied subjects, giving their addresses, the subjects in which they specialise, and a brief description of their mode of operation. By this means it is considered that a notable service will be rendered to seekers after information and research workers by directing them to the nearest institution in a position to supply them promptly with the documentation of which they stand in need. The present work contains a list of most of the important institutes which make a practice of issuing information of the kind indicated, and much credit is due to M. Jean Gerard of Paris, who is responsible for the compiling and editing of the publication.

JONES, T., and T. G. JONES. "*Machine Drawing for the Use of Engineering Students in Science and Technical Schools and Colleges.*" Book I. Revised edition. 8vo, pp. 138. Illustrated. Manchester, 1935: John Heywood, Ltd. (Price 6s.)

The object of this book is to impart to the engineering student a good style of draughtsmanship and a knowledge of the construction of the principal component parts of an engine. It is essential that the student should be trained from the outset in the subject of correct projection, the method adopted throughout the book being that recommended by the British Standards Institution. This method is based on the principles of projection which have always been rigidly applied in the subject of solid geometry. To begin with, a number of general simple details are given which are accompanied by isometric drawings, as the equivalent of models. In this way the student is taught to exercise his imagination and to acquire at an early stage the ability of visualising the shape of an object, or of reading a drawing. In order that the subject may not become a purely mechanical operation, involving no more than neat draughtsmanship, the details of a mechanism are in many cases shown separately, the student being required, with the aid of isometric drawings of reference, to make the necessary views with the parts assembled. Much experience is needed for the dimensioning of a drawing in a correct and efficient manner, and examples of free-hand lettering and figuring are given, for the purpose of practice in providing the workman with all the information and dimensions necessary for the correct shaping of the part. The usefulness of the book will be appreciated by all concerned with the teaching and application of the drawing of mechanical engineering details.

MAIER, A. F. "*Einfluss des Spannungszustandes auf das Formänderungsvermögen der metallischen Werkstoffe.*" 8vo, pp. 47.

Illustrated. Berlin, 1935: V.D.I. Verlag, G.m.b.H. (Price 5 marks.)

This publication consists of a Communication from the Testing Laboratory of the Technical High School, Stuttgart. The author reports the results of experiments made to determine the amount and distribution of the stresses necessary to produce incipient flow in the material and the point at which positive deformation takes place in metallic bodies. The test-specimens were in the form of steel and cast-iron tubes and cylindrical bodies.

SEEGER, G. "*Wirkung von Druckvorspannungen auf die Dauerfestigkeit metallischer Werkstoffe.*" 8vo, pp. 56. Illustrated. Berlin, 1935: V.D.I. Verlag, G.m.b.H. (Price 5 marks.)

This publication consists of a report of an investigation made in the Testing Laboratory of the Technical High School at Stuttgart with the object of determining the effect on metal specimens of alternating stresses of high periodicity and fixed range, under the gradual application, meanwhile, of compressive or tensile stress. Yielding of the material is found to take place at a definite value of the applied mean stress. The work is based, to some extent, on the results obtained by J. H. Smith in his research on the fatigue of metals, as reported in the Journal of the Iron and Steel Institute, 1910, No. II., p. 246.

TAYLOR, F. JOHNSTONE. "*Workshop Practice: A Practical Textbook.*" A revised and enlarged edition of E. Pull's "*Modern Workshop Practice.*" 8vo, pp. 759. With 542 figures in the text. London, 1935: The Technical Press, Ltd. (Price 16s.)

The recent increase in the number of technical institutes and trade schools throughout the United Kingdom has placed within the reach of a large number of students opportunities of systematic instruction in engineering workshop practice. The aim of the author is to meet the requirements of those students by the presentation of a concise treatment of all branches of practical engineering. The work includes chapters on measuring tools and measuring machines, gauge systems and gauges of many different types, workshop tools, cutting speeds and feeds, screw cutting, gears and gear cutting. In addition, the subject of materials and the various methods employed in the melting of cast iron and alloy cast iron are fully dealt with. A description is given of the manufacture of wrought iron, crucible steel, open-hearth steel, alloy steels, and of drop forging and machine forging, and grinding. A number of tables of data useful to engineers is appended, including constants for the conversion of English weights and measures into the corresponding metrical equivalents. With the aid of this book the student or improver should be able to gain an all-round knowledge of the practice and principles of engineering practice.

VOGEL, J. L. F., and W. F. ROWDEN. "*Molybdenum Steels: their Manufacture and Application.*" Prepared for High Speed Steel Alloys, Ltd. With a Foreword by Lord Riverdale of Sheffield. 8vo, pp. 103. Illustrated. Widnes, 1935: High Speed Steel Alloys, Ltd. (Price 5s.)

In recent years the use of molybdenum as an alloying element in steel has developed at a rapid rate, the increase in its use not being due to any distinctive cleansing or deoxidising effect but to the important influence exerted by it in improving the mechanical properties of steel both at normal and at high temperatures. It may be mentioned that Dr. T.

Swinden was one of the first, in 1911, to make a complete study of the effect of adding molybdenum to steel, and its influence on the chemical, mechanical and electrical properties. In the present book is given a short description of the occurrence of molybdenum in the form of a disulphide and of the character of the metal after extraction from its ores. The form in which it is used for alloying with molten steel is a non-volatile compound of molybdenum trioxide and other elements, such as silica and lime, which are not harmful to the product. The metallurgical operations involved in the manufacture of the alloys and their subsequent treatment and properties are all presented in detail from a practical point of view, so that users as well as manufacturers may gain a clear idea of the value of molybdenum as a constituent of high quality steel whether of a simple or complex character.

WARNOCK, F. V. "*Strength of Materials.*" A Text Book Covering the Syllabuses of the B.Sc.(Eng.), A.M.I.C.E. and A.M.I.Mech.E. Examinations in this Subject. Second edition. 8vo, pp. ix + 373. Illustrated. London, 1935: Sir Isaac Pitman and Sons, Ltd. (Price 10s. 6d.)

The work covers the syllabuses of the B.Sc.(Eng.), the A.M.I.C.E. and the A.M.I.Mech.E. examinations in the subject of the strength of materials. It is intended for engineering students, and although its main purpose is to assist those students working for the degrees of the University of London and the examinations of the chief professional engineering institutions, it will be found to cover the courses of other universities, and will, no doubt, prove useful to many practising engineers. In setting forth the derivation of the formulæ essential for the design of structures, beams and machine parts, an enormous number of arithmetical calculations have been made for the estimation of the endless variety of stresses which the material may be called upon to meet in service. The final chapter describes the more important types of testing machines, for measuring the tensile properties, hardness, resistance to impact, and endurance qualities. The methods of cement testing have been brought up to date and are in agreement with the latest specification of the British Standards Institution.

WILLIAMS, R. S., and V. O. HOMERBERG. "*Principles of Metallography.*" Third edition. 8vo, pp. 313. Illustrated. London, 1935: McGraw-Hill Publishing Co., Ltd. (Price 21s.)

This book, the first edition of which was issued in 1919, has now reached its third edition, the text having been rewritten and rearranged in order to take full account of the technical and scientific advances that have since been made in the study of the metallography of both ferrous and non-ferrous metals. As indicated in the preliminary pages, it forms one of the comprehensive International Chemical Series of which Dr. J. F. Norris is the consulting editor. Metallography is one of the most important divisions of the larger field of physical metallurgy and the study of the principles of metallography has done more than anything else to systematise the knowledge of alloys and to make possible the remarkable advances in the use of metals that have taken place in the last thirty years. This volume has been compiled with the object of meeting the needs of those students of general science and engineering who do not have to specialise in metallography, but who must of necessity make use of it in connection with their professional work. In the first chapter the meaning and significance of the alloy diagram is clearly explained. The authors then proceed to discuss the characteristics of non-ferrous alloys of technical importance, with special reference to aluminium alloys, iron and steel alloys, including stainless steel, X-ray methods and magnetic methods of testing, and the optics of metallography. A set of tables is appended containing data of general use to metallographists, together with a descriptive list of the more important books and journals dealing with the subject.

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Printed for THE IRON AND STEEL INSTITUTE *by arrangement with* THE IRON AND COAL TRADES REVIEW *by* F. J. PARSONS, LTD., *London and Hastings.*